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PHYSICAL PROPERTIES OF CASE HARDENED STEELS

By O. W. McMullan

Abstract

This paper consists largely of a number of tables of test data and curves derived therefrom on several of the physical properties of both the case and core of case hardening steels, along with the necessary descriptive matter and the conclusions arrived at from the results. Tests for hardness, strength of bars in bending, of notched and unnotched impact specimens, for distortion, for wear resistance and for resistance of the case against crushing are described. Several of the more common steels in both fine- and coarse-grained sizes with low and high carbon cases and some less common steels were used in making the tests. The influence of case depth was studied by carburizing specimens to from three to four different depths. All samples were carburized in solid compounds. Three heat treatments were given: a direct quench from the carburizing temperature referred to as pot or box quenching, cooling in the box followed by a single reheat, and by a double treatment. The influence of drawing temperature on a few samples is reported. The writer had in mind mainly the suitability of the steels for gears, and the single treatments especially are such as are usually recommended for that purpose. A few test results on medium carbon oil hardening steels are included for comparison with case hardened steels.

Introduction

THERE are being used a considerable number of case hardening steels for gears as well as other case hardened parts. Some plants have standardized on one or two types of steels for gears, while others are using different types of steels for parts under similar service requirements. There has been considerable discussion and disagreement as to the most suitable grain-size to use in any one steel. The same can be said for the carbon content of the case and the depth of the case as well as the method of heat treatment, that is, shall the part be pot-quenched, cooled in the pot and reheated once, or given a double treatment?

A paper presented before the Sixteenth Annual Convention of the Society held in New York City, October 1 to 5, 1934. The author, O. W. McMullan, is chief metallurgist with the Timken-Detroit Axle Co., Detroit. Manuscript received July 5, 1934.

During the last three years, hundreds of tests have been made on several steels and while some have not showed much distinction among the various steels nor pointed to any definite conclusions, others have showed very strikingly, differences both in types of steel and methods of heat treatment. Many tables are listed to show both positive and negative results, so that the reader may arrive at his own conclusions from the data presented. There are naturally some inconsistencies in results, but as many duplicates of tests have been made as time has permitted.

Steels Used

The steels used were six of the most common types of case hardening steel, namely, S.A.E. 1020, 2315, 2512, 3115, 4615 and 6115. It was planned to test both fine-grained and coarse-grained steels of each analysis, except the S.A.E. 6115 which could not be obtained coarse-grained. The two heats each of S.A.E. 2315 and 4615 were very good examples of fine- and coarse-grained heats and the S.A.E. 1020 reasonably good, but there were scattered coarse grains in the fine-grained heat. There was not the desired contrast between the fine- and coarse-grained heats of S.A.E. 2512 and 3115. Both of the S.A.E. 3115 heats were of mixed grain-size. It has been found that those heats having a mixed grain-size might be termed border line heats, that is, if carburized at a slightly lower temperature they will not exhibit grain growth and will be fine-grained, but if carburized higher will develop grain growth and be coarse-grained. This explains disagreements in grain-size if tests on such heats are not carried out under identical conditions. Time as well as temperature has an important influence. In addition to the above eleven heats, a limited number of tests were made on five other analyses. The analysis, grain-size and method of manufacture and key number of the steels are listed in Table I.

Microstructures of Steels

The grain-size and normality of steels are considered by many to have an important influence on their physical properties. McQuaid-Ehn tests were made at 1700 degrees Fahr. on all steels used in this investigation and those on which the majority of the tests were made are shown in Figs. 1 to 11. It will be noticed that some are distinctly normal, while others show more or less thickened and broken

Table 1

Chemical Analyses of Steels Used

No.	Type	Per Cent C	Mn	P	S	Ni	Cr	V	Mo	Si	Grain Size	How Made
1	1020	0.16	0.59	0.029	0.038	....	....	....	....	0.23	2–7	O.H.
T	1020	0.19	0.54	0.017	0.025	....	....	....	....	0.22	2–3	O.H.
2	2315	0.19	0.61	0.012	0.021	3.62	....	....	....	0.23	7	O.H.
U	2315	0.16	0.42	0.016	0.021	3.60	....	....	....	0.18	2–3	O.H.
3	3115	0.18	0.48	0.019	0.031	1.32	0.65	....	....	0.25	2–7	E.F.
W	3115	0.15	0.47	0.022	0.030	1.30	0.59	....	....	0.20	2–6	E.F.
4	4615	0.17	0.54	0.017	0.020	1.77	....	....	0.25	0.29	7–8	E.F.
X	4615	0.16	0.54	0.030	0.020	1.82	....	....	0.24	0.11	3	E.F.
5	2512	0.15	0.53	0.019	0.029	5.00	....	....	....	0.14	6–7	E.F.
Z	2512	0.16	0.45	.....	.....	4.94	....	....	....	....	5	E.F.
6	6115	0.18	0.76	0.019	0.034	....	0.89	0.18	....	0.19	8	E.F.
7	Cr-Ni-Mo	0.16	0.82	.....	.....	1.75	0.80	....	0.40	....	4	Induction
8	Cr-Ni-Mo	0.25	0.67	.....	.....	1.76	0.80	....	0.40	....	4	Induction
K	Krupp	0.08	0.48	.....	....	3.99	1.48	....	....	....	7	E.F.
N	4820	0.18	0.63	.....	.....	3.49	....	....	0.24	....	7	E.F.
	4340	0.39	0.67	.....	.....	1.64	0.61	....	0.33	....	6–7	E.F.
	4640	0.39	0.47	0.032	0.025	1.74	....	....	0.22	0.27	8	E.F.
	5140	0.41	0.67	0.018	0.023	....	1.01	....	....	0.21	3	O.H.

carbide boundaries. They could not be called distinctly abnormal since there is no actual separation of ferrite around the grain boundaries. One steel not shown, S.A.E. 4640, was abnormal. The distinctly normal steels tend to show spines of carbide in the outer part of the case while the less normal, even though fairly coarse-grained, do not do so.

The cores in general follow the grain-size of the cases. Steel No. 1, 1020, although showing a mixed grain-size in the case, was fairly uniform in the core, while No. 3 and W, 3115, were mixed in both the case and the core. Steel 4, 4615, showed a slight mixture of grain sizes in both case and core. Steel 6, 6115, was unusual in that it had a very fine case and badly mixed fine and rather coarse core. A lower carburizing temperature or a short time of carburizing, five hours or less in the furnace, produced a fine-grained uniform core. Steel 4640 also had a very fine case and a mixed core. Steels, 2, 3, 4, U, W, K, and N, showed some banding in the core, but none were extremely banded. Steels 7 and 8 were made in a small induction furnace and cast into ingots four inches square of about 30 pounds each. The ingots were forged into bars 1¼-inch round, but the work was not sufficient to entirely eliminate the cast structure, especially in No. 8.

The microstructures of the heat treated samples were all much as might be expected from the type of steel, grain-size, case depth and heat treatment given. All the samples except some of those of thin case depth and low carbon content, contained austenite when

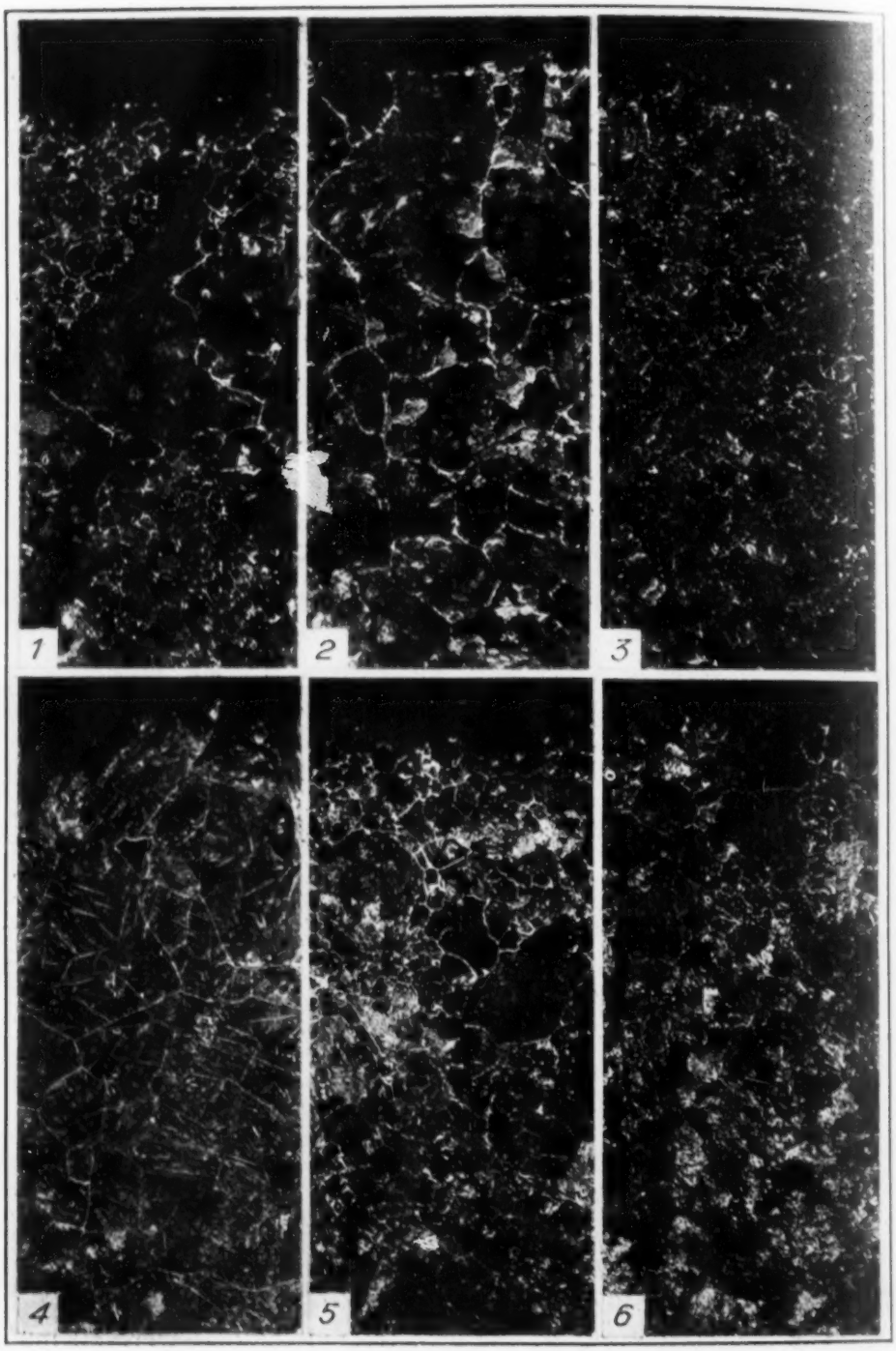

Fig. 1—Steel No. 1, S.A.E. 1020. × 100.
Fig. 2—Steel T, S.A.E. 1020. × 100.
Fig. 3—Steel No. 2, S.A.E. 2315. × 100.
Fig. 4—Steel U, S.A.E. 2315. × 100.
Fig. 5—Steel No. 3, S.A.E. 3115. × 100.
Fig. 6—Steel W, S.A.E. 3115. × 100.

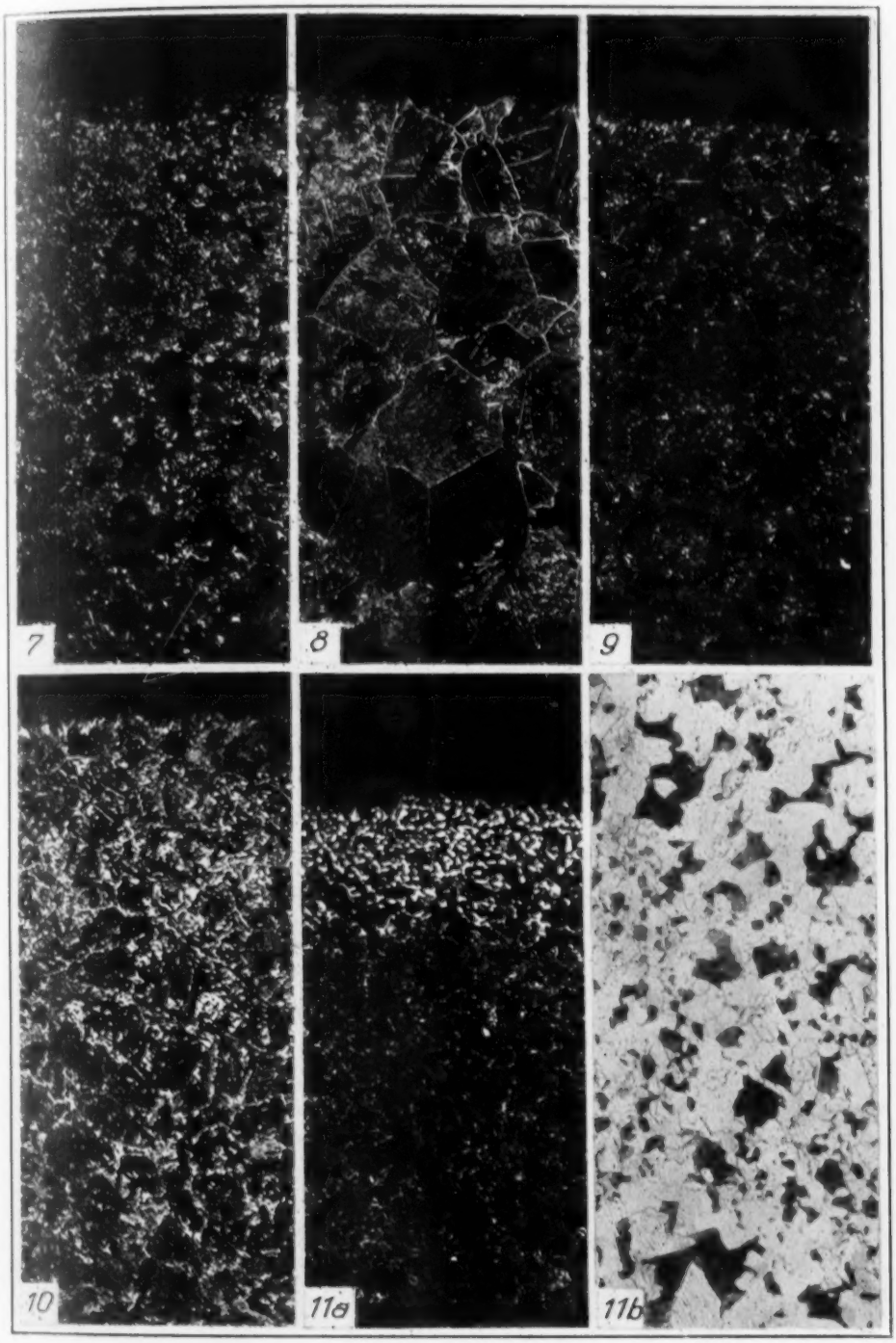

Fig. 7—Steel No. 4, S.A.E. 4615. × 100.
Fig. 8—Steel X, S.A.E. 4615. × 100.
Fig. 9—Steel No. 5, S.A.E. 2512. × 100.
Fig. 10—Steel Z, S.A.E. 2512. The Light Spots were Caused by Too Rapid Cooling and are Martensite and not Ferrite.
Fig. 11a—Steel No. 6, S.A.E. 6115. × 100.
Fig. 11b—Steel No. 6, S.A.E. 6115. × 100. Shows Variation of Core Grain-Size.

quenched direct from the carburizing box. Several of the alloy steels contained austenite in the single treated samples, but none was observed in the double treated samples, but the fine grain of those samples may have made it less distinguishable. The amount of austenite depends upon the alloy content, the carbon content, and quenching temperature, if the cooling rate, etc., is the same. Case depth is a factor, since deeper cases have a deeper border of high carbon content. The S.A.E. 1020 steels (water-quenched) had the least amount of austenite, 3115 and 4615 next and about equal in amount, followed closely by 6115 with a little more. Steels 2320, 4820, 2512 and Krupp contain the most austenite in about the order named. The coarse-grained steels appear to have more austenite than the fine. Comparisons were made on pot-quenched samples, as other hardening temperatures were not identical on all steels. S.A.E. 6115 was unusual among the more common steels since when pot-quenched, its surface layer contained a considerable amount of spheroidized excess carbide while the others showed none. Its presence may be connected with the fact that this steel had a surface layer which seemed to crumble and wear away quite readily under the file. Krupp also showed some carbide in the pot-quenched samples. Figs. 12 and 13 which are of pot-quenched S.A.E. 4615, are typical examples of the difference in appearance of fine- and coarse-grained steels when so treated.

Because of the influence that structure may be considered to have on physical properties, the microstructures of the heat treated samples have been briefly summarized in Tables II to IV.

Tests and Test Specimens

Static tests for case strength were made on bars ¾-inch square and 10 inches long and cased on two opposite sides. The bars were supported 8 inches apart and loaded at the center with the case on the top and bottom. The sides were not case hardened in order to make the cased portion under load more readily determined. The means of preventing case on two sides of the pot-quenched bars was by shaping and grinding 1¼-inch round bars to ¾-inch flats, copper plating, then shaping and grinding the remaining two sides to leave ¾-inch squares. The reheated bars were first made into flats, carburized and cooled in the box, and the remaining two sides finished to make ¾-inch squares before reheating. A few tests were

Table II
Microstructures of Case Hardened Steels

Pot-Quenched Samples

Steel and Grain-Size	Microstructure
1020 Grain 2-7	Case fairly fine, martensite-austenite. Core mixed grain-size, martensitic with troostite and ferrite in the boundaries. No austenite in the low carbon cases and the core is finer. Shallow cased impact specimens have fine core but grain-size increases with case depth due to increased carburizing time. The same is true of the impact cores on most of the other steels.
1020 Grain 2-3	Similar to above but coarse case and core.
2315 Grain 7	Fine austenitic-martensitic case. Core fine martensite-troostite. Amount and depth of austenite in all samples increases with case depth. Less austenite in low carbon cases.
2315 Grain 2-3	Similar but coarse-grain case and core.
3115 Grain 2-7	Fairly fine austenite-martensite case. Core is martensite and troostite variable in grain-size but rather coarse.
3115 Grain 2-6	Same as above.
4615 Grain 7-8	Fine austenitic-martensitic case. Fine troostitic core with coarser grains of martensite.
4615 Grain 3	Same but coarse-grained. Core uniform.
2512 Grain 6-7	Medium austenitic-martensitic case. Much austenite. Fairly fine martensitic core.
2512 Grain 5	Similar to above but coarser case.
6115 Grain 8	Fine austenitic-martensitic case. Excess carbide in case. Medium to rather coarse martensitic-troostitic core in high carbon samples. Low carbon cores are fine-grained since carburizing temperature was too low for grain growth.
Krupp Grain 7	Fairly fine case. Much austenite, some carbide. Martensitic core.
4820 Grain 7	Medium austenitic-martensitic case. Martensite with some troostite in core.
4340 Grain 6-7	Fairly fine austenitic-martensitic case and core.
Cr-Ni-Mo No. 7 & 8 Grain 4	Coarse austenitic-martensitic case. Martensitic core with troostite boundaries in No. 8.

made on bars cased all over and also a few on round bars 7/8-inch in diameter to determine the influence of shape on the results.

Izod impact specimens were the standard 10-millimeter square bars with 45-degree notches two millimeters deep, but made long enough so that two notched and two unnotched tests were made on each specimen. They were cut from 7/8-inch round bars.

Distortion test specimens were designed as shown in Fig. 14. They were cut from bars 1 1/4-inch in diameter. An attempt was made with a simple design to approach the conditions found in a gear, that is, to have a relatively small portion to compare with a gear tooth, extend out from a larger body of metal. The distance between points a and b, also c and d, were measured before carburizing, after cooling in the pot and after final heat treatment, by an amplifying in-

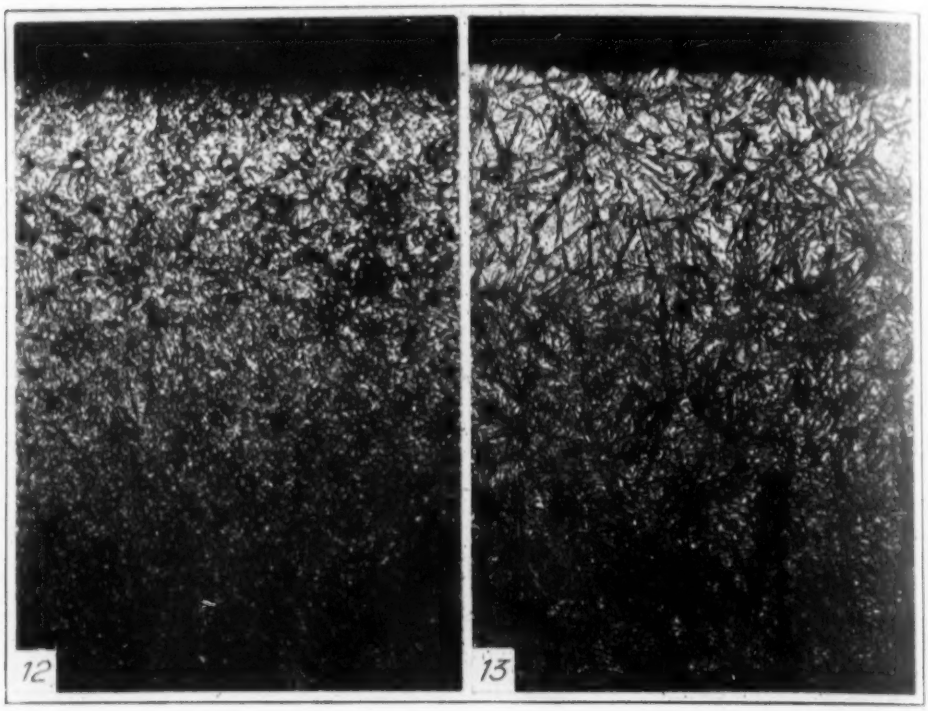

Fig. 12—Case of Fine-Grained S.A.E. 4615 Steel Quenched Directly from the Carburizing Temperature. × 100.
Fig. 13—Case of Coarse-Grained S.A.E. 4615 Steel Quenched Directly from the Carburizing Temperature. × 100.

dicator to the fourth decimal place after removal of scale. This gave the amount of twisting as well as the closing in or spreading out of the prongs.

Case crushing tests were made on the end portions of the bend test specimens not affected by that test. The loads were applied by a special alloy Brinell ball and the diameter of the impressions read, and any sinking or cracking of the case noted. A few checks also were made with hardened rods in place of a Brinell ball.

Wear tests were made on the broken pieces of impact specimens. Those sections between notches were used since they had not been bent in that portion. They were tested on a Timken Roller Bearing Co. lubricant testing machine. This essentially consists of a bearing cup about 2 inches in diameter, rotating at about 500 r.p.m. The specimen is stationary and carried in a lever arm on which weights may be hung. The contact is between the outer face of the cup and the specimen, the length of contact being the width of the bearing or one-half inch.

Standard tensile and Izod test specimens were used for deter-

Table III

Microstructures of Case Hardened Steels

Single Treated Samples

Steel and Grain Size	Microstructure
1020 Grain 2-7	Fine martensite case with increasing network and spheroidized carbide with increase of case depth. Fine core—martensite, troostite and ferrite. The effect of longer carburizing time on coarsening the core grain-size is still plainly visible. The same is true of some of the other steels.
1020 Grain 2-3	Similar but a little coarser-grained.
2315 Grain 7	Case fine martensite with some austenite in deeper cases. Because of partial decarburization during reheating, the austenitic containing layer lies below and not at the surface as in pot-quenched samples. Core fine martensite and troostite.
2315 Grain 2-3	Similar to above but medium grain-size.
3115 Grain 2-7	Fine martensitic case. No visible austenite. Excess carbide network increases with case depth. Core fine martensite and troostite, some ferrite.
3115 Grain 2-6	Similar to above.
4615 Grain 7-8	Fine martensitic case with a little austenite in deeper cases but not in low carbon cases. Fine troostitic core.
4615 Grain 3	Similar to above but case and core a little coarser.
2512 Grain 6-7	Fairly fine austenitic-martensitic case, but little austenite in shallow and low carbon cases. Martensitic core.
2512 Grain 5	Same as above.
6115 Grain 8	Fine martensitic case with spheroidized excess carbide. Trace of austenite in deep case. Fine troostitic core with some ferrite.
Krupp Grain 7	Fine martensitic case and core, not much austenite. Some spheroidized and network excess carbide.
4820 Grain 7	Fine martensite case. Some austenite but no visible excess carbide. Core —martensite and troostite.
4340 Grain 6-7	Fine martensitic case and core. Fine network of excess carbide. Samples drawn at 400 and 800 degrees Fahr. are quite similar in appearance.
Cr-Ni-Mo No. 7 & 8 Grain 4	Fine martensitic case and core. Some spheroidized, some network of carbide. No. 8 shows dendritic outlines and the impact specimen showed an austenitic network independent of the carbide network.

mining core properties. After normalizing, the specimens were packed in sand and heated at the carburizing temperature for 8 hours and quenched from the box or cooled in the box and reheated as desired to correspond to case hardening treatments. They were machined from bars 7/8-inch in diameter.

The values given throughout the paper are in most cases averages from several readings. Over two tons of steel were used in making the laboratory test specimens. An outline of the tests made and the number of duplicate samples are given in Table V. On some specimens, such as for impact and case crushing, several readings were made on each specimen.

The high carbon cases were obtained by ordinary carburizing practice at 1700 degrees Fahr. (925 degrees Cent.). This procedure especially with the deeper cases gives considerable excess carbide

in the case. The low carbon cases were obtained by carburizing in bone meal at 1600 degrees Fahr. (870 degrees Cent.). The intention was to produce cases of eutectoid composition, but the excessive time required for the deep cases resulted in the formation of some excess carbide. Subsequent tables giving zone depths of the case indicate roughly the amount of excess carbide. All carburizing was done in regular production furnaces. Since there was some variation in conditions such as size of container used, amount

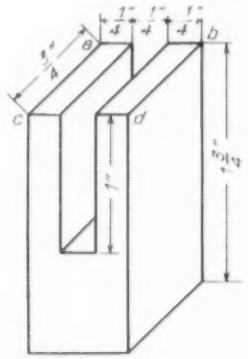

Fig. 14—Sketch of Distortion Test Block Used.

of other work in the furnace, etc., the carburizing times stated cannot in some cases be used as a measure of the rate of carburizing of the various steels. Many samples were run under identical conditions and the observations made on carburizing rates will be mentioned. In all cases the fine and coarse-grained sizes of any one steel were carburized in the same container and, therefore, directly comparable. It is not an easy matter to measure case depths since, for example, if S.A.E. 1020 and 2512 are to be compared at the same cooling rate, the latter even when cooled in the container shows some tendency toward hardening and suppression of ferrite separation and apparently to have a deeper case than true equilibrium conditions would show. The case depth measured is probably a good measure of the effective depth in the finished piece when hardened.

The case depth measurements on the bars are given in Tables VI and VII. By excess zone is meant the depth of case showing excess carbide, the saturated case is the depth to the first appearance of ferrite and includes the excess and eutectoid zones, and the total depth is all the case showing any noticeable increase in carbon content over that of the core. All measurements were made at 100

Table IV
Microstructures of Case Hardened Steels

Double Treated Samples

Steel and Grain-Size	Microstructure
1020 Grain 2-7	Very fine martensitic case. Deeper cases show more spheroidized excess. The surface layer contains no excess carbide because of partial decarburization in this and other steels also. Very fine core of ferrite, martensite and some troostite. The coarsening effect in the core on deeper case depths has been mostly eliminated by the double treatment.
1020 Grain 2-3	Similar to above but a little coarser.
2315 Grain 7	Fine martensitic case. No excess carbide visible. Fine core mostly troostite with some martensite.
2315 Grain 2-3	Similar to above but a little coarser.
3115 Grain 2-7	Fine martensitic case with some network and spheroidized excess carbide. Fine core of martensite and ferrite.
3115 Grain 2-6	Similar to above.
4615 Grain 7-8	Fine martensitic case. Fine troosto-sorbitic core.
4615 Grain 3	Similar. Core fine martensite and troostite, with some ferrite.
2512 Grain 6-7	Fine martensitic case. Fine martensitic troostitic core.
2512 Grain 5	Same as above.
6115 Grain 8	Fine martensitic case with some spheroidized excess carbide beneath the decarburized layer. Fine core of martensite, troostite, and ferrite.
Krupp Grain 7	Fine martensitic case with spheroidized carbide. Martensitic core with patches of troostite.
4820 Grain 7	Fine martensitic case. Martensitic-troostitic core.
4340 Grain 6-7	Fine martensitic case with spheroidized excess carbide. Core martensite and troostite.

Table V
Outline of Laboratory Tests

Specimens For	No. of Case Depths	No. of Heat Treatments on Each	Different Carbon Contents in Case	Duplicates of Each Test	Grain Sizes Each Steel	Total Each Type of Steel
Bend Test	4	3	2	3	2	144
Impact	3	3	2	2	2	72
Warpage	3	3	2	2	2	72
Case Crushing	4	3	2	3	2	144
Wear	Not completed					
Core Tensile		3		2	2	12
Core Impact		3		2	2	12

diameters on samples cooled in the container, the container being removed from the furnace and allowed to cool in air.

The S.A.E. 1020 and 2512 bars were each carburized in a separate box of the same size, while the 2315, 3115, 4615, and 6115 were all carburized together in one larger box and hence under more identical conditions. The smaller container should promote more

Table VI
Case Depth of Pot-Cooled Samples

Sections from ¾" Square Bars for Bend Tests—High Carbon Case

Steel and Grain-Size	Time in Furnace	Excess Zone Inch	Case Depth Saturated Case Inch	Total Case Inch
1020	6 Hours	0.007	0.015	0.034
Grain	11 Hours	0.025	0.039	0.070
2-7	17 Hours	0.028	0.051	0.088
	22 Hours	0.032	0.054	0.090
1020	6 Hours	0.010	0.018	0.035
Grain	11 Hours	0.031	0.041	0.075
2-3	17 Hours	0.033	0.051	0.090
	22 Hours	0.035	0.054	0.103
2315	6⅓ Hours	0.020	0.024	0.039
Grain	11 Hours	0.030	0.038	0.058
7	17 Hours	0.035	0.043	0.075
	22 Hours	0.044	0.054	0.087
2315	6⅓ Hours	0.018	0.022	0.044
Grain	11 Hours	0.030	0.035	0.071
2-3	17 Hours	0.034	0.039	0.078
	22 Hours	0.041	0.045	0.102
3115	6⅓ Hours	0.022	0.031	0.044
Grain	11 Hours	0.031	0.044	0.072
2-7	17 Hours	0.030	0.052	0.087
	22 Hours	0.043	0.067	0.109
3115	6⅓ Hours	0.017	0.028	0.042
Grain	11 Hours	0.029	0.043	0.071
2-6	17 Hours	0.031	0.051	0.084
	22 Hours	0.041	0.068	0.112
4615	6⅓ Hours	0.014	0.018	0.034
Grain	11 Hours	0.025	0.036	0.060
7-8	17 Hours	0.033	0.043	0.071
	22 Hours	0.034	0.048	0.081
4615	6⅓ Hours	0.015	0.024	0.042
Grain	11 Hours	0.030	0.037	0.065
3	17 Hours	0.032	0.041	0.075
	22 Hours	0.038	0.054	0.098
2512	6 Hours	0.015	0.018	0.034
Grain	11 Hours	0.024	0.035	0.058
6-7	17 Hours	0.031	0.044	0.073
	22 Hours	0.034	0.048	0.080
2512	6 Hours	0.015	0.024	0.038
Grain	11 Hours	0.024	0.035	0.061
5	17 Hours	0.030	0.045	0.076
	22 Hours	0.031	0.050	0.078
6115	6⅓ Hours	0.017	0.029	0.044
Grain	11 Hours	0.028	0.048	0.072
8	17 Hours	0.032	0.055	0.093
	22 Hours	0.044	0.074	0.112
Krupp Gr. 7	14 Hours	0.025	0.046	0.068
4820 Gr. 7	14 Hours	0.020	0.042	0.061
.4340 Grain 6-7	14 Hours	0.032	0.061	0.085
Cr-Ni-Mo Grain 4	11 Hours	0.023	0.039	0.058
Cr-Ni-Mo Grain 4	11 Hours	0.024	0.046	0.068

rapid carburizing. The remaining steels were carburized at different times and cannot be compared with the others. Steels 4340, Krupp and 4820, were carburized together in the same container and may be compared as to carburizing rate.

The impact specimens of Tables VIII and IX can be compared

Table VII
Case Depth of Pot-Cooled Samples

Bars Carburized in Bone Meal

Steel and Grain-Size	Time in Furnace	Excess Zone Inch	Case Depth Saturated Case Inch	Total Case Inch
1020 Grain 2-7	16 Hours	0.000	0.006	0.035
	24½ Hours	0.006	0.017	0.054
	34 Hours	0.017	0.035	0.078
1020 Grain 2-3	16 Hours	0.000	0.007	0.037
	24½ Hours	0.007	0.018	0.061
	34 Hours	0.020	0.039	0.088
2315 Grain 7	18½ Hours	Trace	0.008	0.041
	28½ Hours	0.017	0.024	0.058
	36 Hours	0.031	0.037	0.075
2315 Grain 2-3	18½ Hours	Trace	0.008	0.041
	28½ Hours	0.017	0.022	0.065
	36 Hours	0.020	0.028	0.071
3115 Grain 2-7	18½ Hours	0.0.007	0.015	0.044
	28½ Hours	0.018	0.037	0.071
	36 Hours	0.022	0.044	0.082
3115 Grain 2-6	18½ Hours	0.005	0.016	0.044
	28½ Hours	0.020	0.035	0.075
	36 Hours	0.028	0.048	0.082
4615 Grain 7-8	18½ Hours	0.000	0.017	0.041
	28½ Hours	0.014	0.024	0.058
	36 Hours	0.017	0.037	0.075
4615 Grain 3	18½ Hours	Trace	0.013	0.041
	28½ Hours	0.015	0.028	0.068
	36 Hours	0.017	0.037	0.075
2512 Grain 6-7	18½ Hours	Trace	0.017	0.041
	28½ Hours	0.017	0.031	0.058
	36 Hours	0.020	0.034	0.071
2512 Grain 5	18½ Hours	0.000	0.015	0.042
	28½ Hours	0.014	0.031	0.068
	36 Hours	0.020	0.041	0.082
6115 Grain 8	18½ Hours	0.007	0.022	0.048
	28½ Hours	0.020	0.044	0.078
	36 Hours	0.017	0.044	0.078

in much the same way as the bars in Tables VI and VII, except that the 1020 specimens with the high carbon case were run under different furnace conditions than any of the others.

Some previous checks have shown a tendency of coarse-grained steels to have deeper cases than fine-grained, but the results of these tests do not entirely confirm this. There appears to be some tendency, especially with the 2315, to build up as deep or deeper excess zones on the fine-grained steels, but a greater total depth of case on the coarse-grained steels. This may be a result of hindrance of carbon diffusion by grain boundaries. As mentioned, none of the steels used were what might be called abnormal, the tendency being toward thickening and breaking up of carbide boundaries in the fine-grained steels. It was noticed that the S.A.E. 2315 had a very narrow eutectoid zone, in some spots excess carbide extending to almost the beginning of free ferrite. The depth measurements given

Table VIII

Case Depth of Pot-Cooled Samples

Impact Specimens—High Carbon Case

Steel and Grain-Size	Time in Furnace	Case Depth Excess Zone Inch	Case Depth Saturated Case Inch	Total Case Inch
1020 Grain 2-7	4¾ Hours	0.007	0.015	0.041
	3¼ Hours	0.015	0.029	0.059
	11 Hours	0.018	0.031	0.078
1020 Grain 2-3	4¾ Hours	0.012	0.018	0.044
	3¼ Hours	0.018	0.027	0.058
	11 Hours	0.027	0.036	0.076
2315 Grain 7	5 Hours	0.017	0.018	0.041
	6½ Hours	0.024	0.029	0.058
	8 Hours	0.037	0.042	0.078
2315 Grain 2-3	5 Hours	0.015	0.019	0.041
	6½ Hours	0.020	0.024	0.058
	8 Hours	0.035	0.041	0.085
3115 Grain 2-7	5 Hours	0.015	0.022	0.044
	6½ Hours	0.020	0.031	0.058
	8 Hours	0.034	0.051	0.092
3115 Grain 2-6	5 Hours	0.015	0.024	0.044
	6½ Hours	0.020	0.031	0.061
	8 Hours	0.032	0.049	0.086
4615 Grain 7-8	5 Hours	0.014	0.020	0.042
	6½ Hours	0.017	0.024	0.051
	8 Hours	0.034	0.044	0.085
4615 Grain 3	5 Hours	0.015	0.020	0.041
	6½ Hours	0.022	0.032	0.065
	8 Hours	0.034	0.048	0.092
2512 Grain 6-7	5 Hours	0.017	0.022	0.042
	6½ Hours	0.017	0.028	0.058
	8 Hours	0.032	0.044	0.085
2512 Grain 5	5 Hours	0.015	0.020	0.042
	6½ Hours	0.017	0.031	0.061
	8 Hours	0.024	0.041	0.079
6115 Grain 8	5 Hours	0.015	0.025	0.044
	6½ Hours	0.020	0.034	0.061
	8 Hours	0.032	0.052	0.085
Krupp Gr. 7	8 Hours	0.020	0.042	0.061
4820 Gr. 7	8 Hours	0.015	0.030	0.048
4340 Grain 6-7	8 Hours	0.030	0.059	0.078
Cr-Ni-Mo Grain 4	7½ Hours	0.024	0.041	0.058
Cr-Ni-Mo Grain 4	7½ Hours	0.024	0.046	0.065

are in each case on one sample only and comparisons will be made later on hardened case depth measurements. The S.A.E. 6115 appears to be the most rapid to carburize, followed quite closely by 3115, then 1020, 2315, with 4615 and 2512 the slowest and at about the same rate.

Heat Treatment of the Steels

All samples were drawn 1 hour at 325 degrees Fahr. in a Homo furnace unless otherwise stated later in tables of results. Samples carburized in bone meal were, in case of pot-quenched sam-

Table IX
Case Depth of Pot-Cooled Samples

Impact Specimens Carburized in Bone Meal

Steel and Grain-Size	Time in Furnace	Excess Zone Inch	Case Depth Saturated Case Inch	Total Case Inch
1020	11 Hours	0.000	0.003	0.031
Grain	22 Hours	0.003	0.014	0.048
2-7	31 Hours	0.014	0.032	0.065
1020	11 Hours	0.000	0.005	0.032
Grain	22 Hours	0–0.007	0.014–0.020	0.051
2-3	31 Hours	0.000	0.028	0.075
2315	13½ Hours	0.000	0.007	0.034
Grain	23½ Hours	0.007	0.017	0.048
7	31 Hours	0.031	0.034	0.068
2315	13½ Hours	0.000	0.005	0.034
Grain	23½ Hours	0.010	0.014	0.054
2-3	31 Hours	0.028	0.031	0.075
3115	13½ Hours	0.000	0.008	0.037
Grain	23½ Hours	0.010	0.026	0.058
2-7	31 Hours	0.024	0.051	0.092
3115	13½ Hours	0.000	0.006	0.033
Grain	23½ Hours	0.012	0.024	0.058
2-6	31 Hours	0.026	0.052	0.092
4615	13½ Hours	0.000	0.006	0.034
Grain	23½ Hours	0.007	0.017	0.048
7-8	31 Hours	0.017	0.034	0.068
4615	13½ Hours	0.000	0.012	0.037
Grain	23½ Hours	0.007	0.010	0.054
3	31 Hours	0.024	0.031	0.075
2512	13½ Hours	0.000	0.008	0.034
Grain	23½ Hours	0.007	0.017	0.048
6-7	31 Hours	0.024	0.034	0.068
2512	13½ Hours	0.000	0.010	0.035
Grain	23½ Hours	.....	0.020	0.054
5	31 Hours	0.017	0.034	0.075
6115	13½ Hours	0–0.003	0.016	0.038
Grain	23½ Hours	0.014	0.031	0.058
8	31 Hours	0.028	0.061	0.092

Table X
Heat Treatments of Carburized Steels

Steel No.	Type	Pot Quench Degrees Fahr.	Single Treatment Degrees Fahr.	Double Treatment Degrees Fahr.	
1 and T	1020	1700-Water	1425-Water	1625-Oil	1425-Water
2 and U	2315	1700-Oil	1475-Oil	1550-Oil	1380-Oil
3 and W	3115	1700-Oil	1475-Oil	1550-Oil	1425-Oil
4 and X	4615	1700-Oil	1500-Oil	1550-Oil	1400-Oil
5 and Z	2512	1700-Oil	1450-Oil	1550-Oil	1350-Oil
6	6115	1700-Oil	1625-Oil	1650-Oil	1475-Oil
7 and 8	C-N-M	1700-Oil	1500-Oil		
	Krupp	1700-Oil	1450-Oil	1550-Oil	1400-Oil
	4820	1700-Oil	1475-Oil	1550-Oil	1400-Oil
	4340	1700-Oil	1525-Oil	Pot Quench	1400-Oil

ples, quenched from the carburizing temperature of 1600 instead of 1700 degrees Fahr. as listed in Table X.

Results From Bend Tests

The bars ¾-inch square were cased on the top and bottom

sides only, supported 8 inches apart and loaded at the center. Deflection readings were taken at the center for every 200 pounds of load applied. It was at first attempted to determine the occurrence of the first crack by hearing, but this was found impossible in some of the alloy steels. It was necessary to use an oblique light and watch for cracks to appear. They could be detected quite readily on the surface as cleaned after heat treatment. In many cases they started slowly at one edge or the other, sometimes in the center of the bar, but many times also they occurred with a click across the entire bar. The first crack was recorded because it is believed that the life of a case hardened part would be short after the case had once cracked. Except for the thinner cases in some of the alloy steels, the first crack was at or very close to the elastic limit. The stress and deflection at the first crack of the case, and at the maximum load are recorded in the tables. The breaking load and maximum deflection were not always identical with those at the maximum load. The stress in the extreme fibers has been calculated by the usual formula for tension in bending.

$$S = \frac{Mc}{I}$$

The appearance of the fracture of both the case and core, the case depth, Rockwell hardness of the case, and core Brinell are recorded in Tables XI to XVI, along with the load and deflection readings for each type of steel and grain-size. Tables XI to XIII are on the high carbon bars and Tables XIV to XVI are on the low carbon cases.

The reheated bars were all treated in an electric furnace which produced some scale and decarburization, the double treated samples being most affected especially because of the higher first heat. Scale was removed by shot blasting. The reheating produced in most cases a thin skin that could be touched with a file, especially on the double treated samples. This was especially true of the S.A.E. 6115 because of the high hardening temperatures used and perhaps it would have been better to have compared this steel by heating in cyanide according to a quite common procedure. The file hardness of the various steels will be compared on the impact specimens, since the smaller sections, because of the shorter time in the furnace, were less influenced by furnace conditions. Also a smoky furnace atmos-

phere was produced when treating the impact specimens by placing carbonaceous material in the furnace. In general the file hardness of the larger samples was in the same order as the smaller ones. The Rockwell hardness of the case is dependent upon the amount of martensite present. The combination of alloying elements, high carbon and high quenching temperatures produce austenite, which lowers the hardness according to the amount present. Some decrease in Rockwell hardness resulting from decarburization occurred in reheated samples, especially where the case was low in carbon because of being thin or of low carbon content. High heat treating temperatures, such as used for 6115, promoted decarburization. There were some soft spots in the plain carbon steel, probably due to steam pockets and low hardening capacity. The higher Rockwell readings represent what the case would be without decarburization and the low readings indicate the extent of decarburization on the surface.

The number of cracks shown in a case hardened part when broken is frequently considered as a measure of its shock resistance. The results from the load values in the bend test and a comparison of impact values on pieces of similar treatment and case depth show this is not true. There is apparently no relationship between the number of cracks and shock resistance, except that arising from differences in case depth. While the impact values on single treated samples are lowest in both cases, the S.A.E. 1020 steel showed the least cracks when single treated and the S.A.E. 3115 the most. The deep cases always cracked less than the shallow cases, the coarse-grained steels cracked more than the fine-grained steels, and in general the pot-quenched samples had more cracks in the case than did the single or double treated samples. The numbers in Table XVII give the minimum to maximum number of cracks in duplicate samples. The number of cracks in each steel are listed vertically downward in order of increasing case depth.

Discussion of Results From Bend Tests

Woodvine[1] has discussed the fatigue range in a carburized specimen and originated the diagram shown in Fig. 15 to illustrate the stresses. The vertical line through O represents the neutral axis of zero stress in a beam, the point A as the stress at the surface of the case. Stresses between the center and surface would lie along

[1] J. G. R. Woodvine, "The Behavior of Case Hardened Parts Under Fatigue Stresses," The Iron and Steel Institute, Carnegie Scholarship Memoirs, Vol. XIII, 1924.

Table XI

Bend Tests—High Carbon Case—Pot Quench

Steel No. and Grain-Size	Case Strength First Crack Pounds per Square Inch	Deflection First Crack Inch	Max. Case Strength Pounds per Square Inch	Maximum Deflection Inch	Case Depth Inch	Rockwell of Case	Core Brinell	Fracture
1020	155,500	.086	193,000	.245	.030	62-63	217	Fine case. Silky core.
Grain-Size	158,300	.077	200,300	.126	.045	60-62	255	
2-7	143,700	.066	177,560	.099	.065	60-62	262	Case mostly fine. Core mostly silky.
	146,300	.072	164,400	.068	.072	60-63	255	
1020	138,500	.060	203,100	.116	.034	61-63	293	Coarse case. Core ½ fairly fine crystalline, rest silky.
Grain-Size	113,000	.048	202,600	.133	.048	60-62	285	
2-3	95,700	.036	175,700	.103	.064	61-63	277	
	113,000	.055	172,500	.104	.074	61-62	285	
2315	258,000	.159	287,100	.222	.035	57-59	402	Fine case. Silky core.
Grain-Size	205,000	.112	213,300	.122	.054	52-54	402	
7	219,000	.122	219,300	.122	.064	55-57	415	
	210,700	.124	210,700	.124	.072	52-54	430	
2315	212,000	.121	289,000	.374	.035	57-59	321	Coarse case. Silky core.
Grain-Size	180,000	.099	270,500	.290	.045	53-56	331	
2-3	165,000	.090	239,900	.184	.062	55-58	321	
	164,000	.085	225,400	.174	.072	53-55	341	
3115	187,900	.099	227,000	.173	.030	57-61	302	Case fairly fine. Core silky.
Grain-Size	178,900	.092	208,800	.151	.048	54-57	302	
2-7	165,000	.080	169,200	.086	.066	57-59	311	Medium case. Core mostly silky.
	165,800	.083	169,600	.090	.080	55-56	321	
3115	174,400	.099	231,000	.342	.030	56-62	269	Fairly fine case. Core mostly silky.
Grain-Size	158,400	.083	227,600	.332	.046	56-59	262-269	
2-6	155,500	.078	205,900	.228	.064	57-59	285	
	155,000	.077	202,300	.238	.076	56-57	277	
4615	192,700	.122	241,600	.350	.027	59-62	302	Fine case. Silky core.
Grain-Size	188,000	.105	231,700	.298	.045	57-59	311	
7-8	191,700	.105	213,200	.238	.060	60	269	
	203,200	.110	203,200	.109	.078	55-59	311	
4615	179,000	.092	261,700	.418	.028	56-62	277	Coarse case. Core mostly silky.
Grain-Size	148,000	.074	251,900	.353	.045	56-58	277	
3	149,700	.075	240,900	.323	.066	56-59	293	
	145,000	.075	228,300	.302	.074	56-57	269	
2512	253,300	.162	315,500	.414	.031	54-57	415	Case fairly fine to medium. Silky core.
Grain-Size	225,600	.150	286,400	.362	.045	47-51	402	
6-7	203,600	.115	231,400	.163	.060	52-53	415	
	176,300	.096	190,300	.122	.075	50-52	402	

Table XI (Continued)

2512	253,000	.161	266,100	.378	.032	53-56	388	Medium case. Silky core.
Grain-Size	203,000	.126	277,300	.351	.048	49-51	388	
5	187,800	.104	236,600	.198	.060	51-54	375	

Table XI (Continued)

2512	253,000	.161	266,100	.378	.032	53-56	388	Medium case. Silky core.
Grain-Size	203,000	.126	277,300	.351	.048	49-51	388	
5	187,800	.104	236,600	.198	.060	51-54	375	
	193,000	.114	212,200	.151	.078	48-51	388	
6115	191,200	.114	255,700	.450	.028	56-60	302	Very fine case. Silky core.
Grain-Size	172,200	.090	243,900	.288	.050	54-56	302	Fine case. Core ½ fine cryst.
8	170,900	.082	226,500	.234	.066	58-60	302	Fairly fine case. Core mostly cryst.
	158,400	.080	209,100	.227	.080	55-57	293	
Krupp Grain-Size 7	209,700	.123	267,900	.358	.050	41-45	363	Case rather coarse. Silky core.
4820 Grain-Size 7	194,400	.097	220,200	.126	.055	55-58	415	Fine case. Silky core.
4340	196,250	.092	196,400	.092	.054	52-56	514	400° Draw. Case and core both fairly fine cryst.
Grain-Size 6-7	347,000	.215	347,000	.215	.060	51-53	430	800° Draw. Coarser than 400° draw.
Cr-Ni-Mo Grain-Size 4	202,000	.108	290,300	.325	.050	52-55	415	Rather coarse case. Silky core.
Cr-Ni-Mo Grain-Size 4	209,100	.108	272,500	.183	.050	53-54	495	Rather coarse case. Silky core.

Table XII

Bend Tests—High Carbon Case—Single Treated

Steel No. and Grain Size	Case Strength First Crack Pounds Per Square Inch	Def. 1st Crack Inch	Maximum Case Str. Pounds Per Square Inch	Max. Def. Inch	Case Depth Inch	Rockwell of Case	Core Brin.	Fracture
1020	137,800	0.090	147,900	0.151	0.030	63-66	179	Fine case. Core mostly fairly fine cryst.
Grain Size	137,000	0.074	142,000	0.081	0.055	61-65	187	Fine case. Rather coarse cryst. core.
2-7	163,000	0.090	163,200	0.089	0.070	60-66	192	Fine to fairly fine case. Coarse cryst. core.
	157,000	0.079	156,800	0.079	0.078	62-66	179	Fine to fairly fine case. Coarse cryst. core.
1020	158,000	0.094	159,800	0.104	0.030	63-66	196	Fine to fairly fine case. Core mostly med. cryst.
Grain Size	137,700	0.066	139,400	0.071	0.058	64-66	192	Fine to fairly fine case. Core medium to rather coarse, cryst.
2-3	166,000	0.084	166,100	0.084	0.075	64-66	202	
	146,000	0.069	146,000	0.069	0.082	62-66	187	
2315	226,700	0.134	240,700	0.176	0.040	57-60	340	Fine case. Silky core.
Grain Size	228,300	0.130	233,600	0.155	0.055	56-58	364	
7	252,000	0.163	252,100	0.163	0.064	56-58	388	
	222,000	0.123	222,400	0.125	0.078	55-57	364	
2315	182,300	0.102	222,600	0.218	0.038	58-60	269	Medium to rather coarse case. Silky core.
Grain Size	167,000	0.085	202,800	0.190	0.058	57-59	269	
2-3	173,000	0.089	196,800	0.202	0.064	58-59	248	
	155,500	0.076	178,300	0.162	0.078	57-59	269	

This table is continued on the next page.

Table XII (Continued)

3115	164,200	0.082	228,600	0.253	0.038	62-63	277	Fine case. Silky core.
Grain Size	148,700	0.071	222,700	0.268	0.058	62-64	302	
2-7	167,500	0.083	213,600	0.222	0.065	60-63	302	
	185,000	0.092	192,900	0.193	0.078	61-64	269	
3115	140,100	0.074	205,200	0.400	0.036	61-63	235	Fine case. Silky core.
Grain Size	152,800	0.080	193,900	0.343	0.052	60-64	235	
2-6	162,400	0.085	194,500	0.283	0.065	61-63	228	
	165,500	0.083	190,000	0.292	0.080	59-63	235	
4615	168,000	0.100	216,250	0.408	0.038	59-62	255	Fine case. Silky core.
Grain Size	182,000	0.107	204,900	0.300	0.048	59-62	229	
7-8	189,000	0.100	216,600	0.317	0.064	61-62	262	
	199,000	0.111	207,600	0.275	0.075	60-61	235	
4615	148,000	0.075	234,700	0.440	0.036	61-62	262	Fairly fine to medium case. Core ⅓ to ½ fine cryst., rest silky.
Grain Size	129,000	0.060	216,300	0.400	0.053	60-62	241	
3	136,000	0.065	211,900	0.400	0.060	60-61	241	
	137,300	0.063	203,700	0.325	0.080	60-62	255	
2512	275,100	0.184	304,000	0.263	0.032	57-59	388	Fine case. Silky core.
Grain Size	237,100	0.148	252,700	0.164	0.050	54-57	388	
6-7	258,300	0.173	258,300	0.173	0.070	54-57	388	
	239,600	0.139	239,600	0.139	0.080	54-56	375	
2512	242,800	0.146	289,500	0.272	0.028	56-60	415	Fine case. Silky core.
Grain Size	257,400	0.159	257,400	0.163	0.055	55-57	388	
5	250,100	0.148	250,100	0.148	0.065	54-57	402	
	244,100	0.146	244,100	0.146	0.075	55-57	364	
6115	200,400	0.115	246,100	0.370	0.039	59-64	286	Very fine case. Cores vary from all silky to part fine cryst. Reedy.
Grain Size	210,700	0.129	213,600	0.137	0.056	58-63	262	
8	192,800	0.105	239,600	0.224	0.065	53-59	262	
	208,500	0.113	215,600	0.169	0.080	56-62	262	
Krupp Grain Size-7	295,100	0.176	297,400	0.176	0.050	61-63	415	Fine to fairly fine case. Fine core.
4820 Grain Size-7	240,600	0.133	256,650	0.160	0.052	61-62	401	Fine case. Silky core.
4340	337,100	0.177	337,100	0.177	0.054	60-62	514	400 degrees Draw. Fine case and core.
Grain Size	386,900	0.361	386,900	0.361	0.060	52-53	430	800 degrees Draw. Coarser than 400 degrees draw.
6-7								
Cr-Ni-Mo Grain Size-4	231,900	0.127	277,200	0.279	0.050	62-63	364	Fine case. Silky core.
Cr-Ni-Mo Grain Size-4	256,700	0.139	256,750	0.140	0.050	62-63	477	Fine case. Silky core.

Table XIII
Bend Tests—High Carbon Case—Double Treated

Table XIII

Bend Tests—High Carbon Case—Double Treated

Steel No. and Grain-Size	Case Strength First Crack Pounds per Square Inch	Deflection First Crack Inch	Max. Case Strength Pounds per Square Inch	Maximum Deflection Inch	Case Depth Inch	Rockwell of Case	Core Brinell	Fracture
1020	133,000	.090	159,400	.258	.030	59-66	179	Fine case. Core ½ or more silky, rest fine cryst.
Grain-Size	147,000	.092	164,000	.292	.055	56-66	179	
2-7	155,700	.089	163,300	.180	.066	56-66	179	
1020	143,700	.095	162,900	.188	.037	59-64	192	Fine case. Core mostly fairly fine cryst.
Grain-Size	131,700	.068	168,000	.243	.062	58-66	192	Eutectoid zone tends to show as coarse band.
2-3	165,500	.090	166,400	.111	.072	58-66	187	
2315	260,000	.163	262,600	.170	.038	58-60	340	Fine case. Silky core.
Grain-Size	257,300	.170	259,300	.190	.064	56-60	321	
7	296,000	.233	295,900	.233	.068	56-60	340	
2315	227,500	.153	273,900	.196	.035	59-60	241	Fine case. Silky core.
Grain-Size	235,700	.178	247,200	.245	.055	58-60	235	
2-3	236,000	.163	236,100	.188	.065	59-60	241	
3115	179,200	.114	196,300	.187	.038	60-63	262	Fine case. Silky core.
Grain-Size	192,000	.114	193,300	.141	.052	55-63	269	
2-7	192,700	.109	192,700	.128	.065	59-63	248	
3115	132,400	.086	161,800	.343	.032	58-61	196	Fine case. Silky core.
Grain-Size	164,900	.114	165,700	.229	.048	55-62	196	
2-6	157,800	.093	167,700	.250	.060	55-61	207	
4615	173,400	.114	196,700	.352	.035	60-63	248	Fine case. Silky core.
Grain-Size	180,000	.125	193,500	.223	.048	57-62	207	
7-8	200,000	.130	209,000	.253	.064	57-63	223	
4615	155,300	.090	193,300	.300	.040	62-63	235	Fine case. Silky core.
Grain-Size	177,000	.104	194,900	.174	.050	62-63	228	
3	174,000	.098	184,500	.266	.065	60-63	235	
2512	303,000	.204	307,300	.220	.036	56-59	401	Fine case. Silky core.
Grain-Size	296,900	.210	296,900	.210	.052	56-57	375	
6-7	311,100	.216	311,100	.213	.070	55-58	363	
2512	254,500	.158	274,200	.206	.027	57-60	364	Fine case. Silky core.
Grain-Size	314,700	.235	314,400	.235	.048	57-58	375	
5	298,000	.202	298,200	.202	.065	56-59	364	
6115	170,900	.128	182,000	.197	.034	57-62	235	Very fine case. Silky core.
Grain-Size	170,100	.096	179,900	.162	.056	54-61	235	
8	196,000	.119	205,200	.158	.060	52-63	248	
Krupp Grain-Size-7	325,700	.215	325,700	.215	.050	58-62	388	Fine case. Silky core.
4820 Grain-Size-7	345,200	.294	345,200	.294	.055	59-62	388	Fine case. Silky core.
4340	228,500	.106	228,500	.105	.054	63-64	514	400° Draw. Fine case and core.
Grain-Size	352,700	.200	352,700	.200	.060	53-54	444	800° Draw. Coarser than 400° draw
6-7								

the line OA. BC represents the fatigue limit of the case and DG the fatigue limit of the core. The line BPQG therefore represents the fatigue limit of the carburized specimen. Since the stress line OA lies to the left of BPQG, the fatigue limit is not exceeded by such a load. Had the core fatigue limit been lower, shifting DG to the left so that point Q would be to the left of the line OA, fatigue failure would take place in the core just below the case. The strength of a cased part, provided the fatigue limit of the case is sufficiently

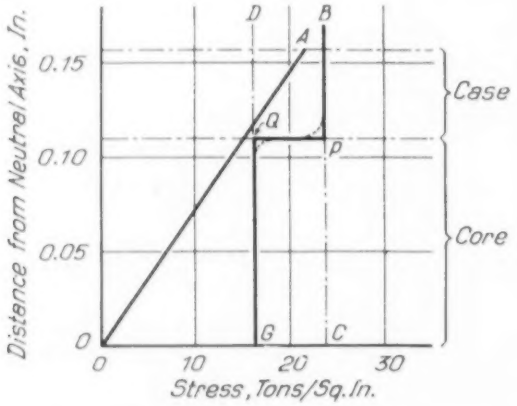

Fig. 15—Diagram Indicating the Fatigue Range in Carburized Steels. (After Woodvine.)

high, can therefore be increased by either increasing the core hardness and strength, or increasing the case depth, that is, shifting point Q to the right or lowering it to again bring it to the right of OA. The diagram was used to illustrate the stresses in a normalized specimen. Mr. Woodvine mentioned the effect of internal stresses set up in hardening, but thought it probable the same relationship would hold in a hardened part.

The same diagram can be used to illustrate stresses in a beam and it might be expected that increased case depth and increased core hardness would give increased load carrying capacity. The results of this investigation, however, show more exceptions than agreements with this assumption. All the pot-quenched samples except for the first crack in the fine-grained S.A.E. 4615, decrease in strength with increase in case depth when carburized according to standard practice, and this decrease is very pronounced in many cases. The majority of the single treated samples likewise decrease in strength with increase of case depth, especially as to maximum load and more so in the coarse- than in the fine-grained steels. Nearly

Table XIV

Bend Tests—Low Carbon Case—Pot Quench

Steel No. and Grain-Size	Case Strength First Crack Pounds per Square Inch	Deflection First Crack Inch	Max. Case Strength Pounds per Square Inch	Maximum Deflection Inch	Case Depth Inch	Rockwell of Case	Core Brinell	Fracture
1020	150,250	.107	166,300	.275	.024	58-64	187	Fine case. Core ½ silky, rest fine cryst.
Grain-Size	176,700	.114	181,650	.186	.042	63.5	196	Fine case. Core silky.
2-7	202,800	.096	202,400	.096	.075	62-65	277	Fine case. Core mostly silky but uneven.
1020	187,700	.095	256,500	.171	.025	63-64	286	Rather Coarse Case. Core ½ or more fine cryst.
Grain-Size	142,100	.062	208,300	.115	.044	64-65	277	Rest silky.
2-3	118,200	.044	164,200	.085	.070	62-64	286	
2315	295,800	.215	302,900	.264	.027	60	364	Fine case. Silky core.
Grain-Size	282,650	.181	282,650	.181	.045	60	388	
7	289,500	.178	289,500	.178	.075	56-58	388	
2315	241,850	.159	285,350	.410	.030	60-61	293	Rather coarse case. Silky core.
Grain-Size	227,300	.127	266,850	.225	.050	60-61	311	
2-3	203,200	.103	216,500	.120	.075	58-61	332	
3115	215,600	.138	254,100	.350	.023	60-62	286	Fairly fine case. Core mostly silky, rest fine cryst.
Grain-Size	227,650	.128	235,700	.196	.045	62-63	302	
2-7	224,100	.118	224,100	.118	.070	60-62	311	
3115	215,650	.173	226,950	.428	.024	61-62	241	Fairly fine case. Core mostly silky.
Grain-Size	195,450	.108	234,600	.270	.042	63	277	
2-6	183,350	.094	196,500	.190	.073	61-63	262	
4615	214,600	.150	242,400	.538	.024	61-63	311	Fine case. Silky core.
Grain-Size	223,900	.140	242,700	.295	.040	63	262	
7-8	221,250	.122	221,250	.122	.078	61-63	311	
4615	193,450	.110	257,250	.445	.025	61-63	255	Fairly fine case. Core ⅓ fairly fine cryst., rest silky.
Grain-Size	190,500	.101	247,650	.365	.048	62-63	248	Medium case. Core mostly silky.
3	165,000	.079	223,300	.277	.070	60-62	277	Rather coarse case. Core ⅓ fairly fine cryst.
2512	302,500	.234	330,000	.450	.023	58	415	Fine case. Silky core.
Grain-Size	307,300	.220	307,400	.230	.046	57-59	415	
6-7	288,850	.186	288,850	.186	.072	55-56	415	Case somewhat coarser.
2512	274,950	.198	306,700	.450	.028	57-59	364	Fairly fine case. Silky core.
Grain-Size	237,350	.144	266,000	.202	.040	58-59	388	Medium case. Silky core.
5	229,650	.129	229,650	.129	.068	55-57	388	Rather coarse case. Silky core.
6115	249,750	.297	249,750	.297	.024	59-64	262	Very fine case. Core ½ fine cryst., rest silky.
Grain-Size	233,700	.154	257,000	.188	.044	62-63	293	Fine case. Core ⅓ fine cryst., reedy.
8	259,100	.162	259,100	.162	.072	60-62	293	Fine case. Fine cryst. core.

Table XV

Bend Tests—Low Carbon Case—Single Treated

Steel No. and Grain-Size	Case Strength First Crack Pounds per Square Inch	Deflection First Crack Inch	Max. Case Strength Pounds per Square Inch	Maximum Deflection Inch	Case Depth Inch	Rockwell of Case	Core Brinell	Fracture
1020	144,400	.103	153,250	.200	.025	63-64	179	Fine case. Core ⅓ silky, rest fine cryst.
Grain-Size	151,900	.093	157,400	.170	.045	63-65	187	Fine case. Core mostly fine cryst.
2-7	140,350	.068	140,350	.068	.075	63-66	183	Fine case. Coarse crystalline core.
1020	165,500	.108	170,700	.116	.025	62-65	196	Fairly fine case. Core fairly fine, mostly cryst.
Grain-Size	138,200	.077	138,200	.077	.046	63-66	207	Fairly fine case. Core rather coarse, cryst.
2-3	175,750	.089	175,750	.089	.080	62-65	207	Fairly fine case. Core rather coarse, cryst.
2315	283,000	.229	283,000	.229	.025	58-60	351	Fine case. Silky core.
Grain-Size	257,700	.148	257,700	.148	.048	58-61	375	
7	245,900	.136	245,800	.149	.075	58-60	340	
2315	221,500	.145	228,750	.180	.026	58-60	255	Medium case. Silky core.
Grain-Size	211,950	.115	215,300	.135	.055	59-60	286	
2-3	171,650	.084	178,400	.098	.075	58-60	269	Rather coarse case. Silky core.
3115	179,500	.108	212,100	.262	.028	61-64	269	Fine case. Silky core.
Grain-Size	159,500	.078	222,750	.192	.050	62-64	269	
2-7	190,350	.098	199,600	.132	.072	62-63	269	
3115	155,200	.106	189,300	.388	.025	58-61	223	Fine case. Silky core.
Grain-Size	165,000	.097	197,400	.292	.044	59-62	235	
2-6	176,200	.092	197,100	.268	.064	62-64	235	
4615	178,850	.166	199,500	.432	.026	56-60	255	Fine case. Silky core.
Grain-Size	190,000	.118	208,400	.235	.050	60-63	269	
7-8	217,600	.120	221,250	.225	.065	59-63	269	
4615	169,200	.110	199,900	.525	.024	58-61	235	Fine case. Silky core.
Grain-Size	159,500	.085	207,100	.437	.040	62-63	228	Fine case. Core ¼ fine crystalline.
3	141,600	.066	204,600	.388	.068	61-63	241	Fairly fine case. Core ¼ fine cryst.
2512	300,100	.216	311,600	.248	.030	57-60	388	Fine case. Silky core.
Grain-Size	286,000	.180	293,800	.197	.050	57-59	401	
6-7	290,000	.176	290,000	.176	.070	57-58	388	
2512	273,200	.184	288,400	.220	.025	57-59	388	Fine case. Silky core.
Grain-Size	255,200	.149	257,400	.159	.045	58-60	388	
5	216,000	.116	223,200	.125	.070	57-59	364	
6115	209,300	.134	232,600	.345	.031	61-64	277	Very fine case. Silky, reedy core.
Grain-Size	194,000	.103	235,400	.202	.050	56-63	277	Very fine case. Core mostly silky, reedy.
8	221,400	.121	255,900	.184	.078	57-62	286	Very fine case. Core ½ fine cryst.

Table XVI

Table XVI

Bend Tests—Low Carbon Case—Double Treated

Steel No. and Grain-Size	Case Strength First Crack Pounds per Square Inch	Deflection First Crack Inch	Max. Case Strength Pounds per Square Inch	Maximum Deflection Inch	Case Depth Inch	Rockwell of Case	Core Brinell	Fracture
1020	136,750	.103	150,750	.225	.022	53-54	183	Fine case. Core mostly silky.
Grain-Size	141,200	.091	159,600	.270	.042	59-65	179	
2-7	145,100	.076	156,500	.300	.072	58-65	174	
1020	154,700	.113	160,300	.147	.025	58-62	187	Fine case. Core mostly fairly fine crystalline.
Grain-Size	149,300	.089	165,400	.262	.045	61-65	187	
2-3	146,200	.072	149,500	.089	.080	60-64	212	
2315	310,100	.299	310,800	.301	.030	56-60	321	Fine case. Silky core.
Grain-Size	304,500	.213	304,500	.213	.060	57-60	352	
7	273,200	.161	273,200	.161	.075	57-59	340	
2315	241,600	.205	241,600	.244	.030	58-60	241	Fine case. Silky core.
Grain-Size	256,000	.185	256,000	.185	.047	59-61	241	
2-3	278,500	.192	278,500	.192	.068	59-60	248	
3115	164,600	.108	182,300	.182	.028	59-61	269	Fine case. Silky core.
Grain-Size	195,400	.121	195,400	.121	.048	59-62	255	
2-7	197,300	.108	197,300	.108	.068	58-62	262	
3115	146,000	.149	157,500	.400	.025	55-59	202	Fine case. Silky core.
Grain-Size	156,900	.117	165,800	.222	.044	59-62	192	
2-6	178,000	.107	178,000	.107	.065	60-62	187	
4615	182,700	.165	188,100	.248	.030	58-62	241	Fine case. Silky core.
Grain-Size	196,250	.148	199,400	.219	.055	60-63	235	
7-8	217,300	.148	217,300	.178	.073	60-62	241	
4615	166,000	.116	188,600	.275	.027	59-61	235	Fine case. Silky core.
Grain-Size	180,500	.111	190,500	.181	.050	61-63	223	
3	166,850	.087	188,400	.163	.075	62	241	
2512	317,250	.288	321,450	.288	.030	56-59	363	Fine case. Silky core.
Grain-Size	334,250	.236	334,250	.236	.050	56-59	401	
6-7	293,100	.167	293,100	.167	.072	56-59	401	
2512	315,700	.267	318,700	.279	.028	56-58	388	Fine case. Silky core.
Grain-Size	290,700	.193	290,700	.193	.045	57-59	364	
5	284,000	.168	284,000	.168	.068	58-59	388	
6115	136,300	.104	155,400	.197	.030	50-61	228	Very fine case. Silky core.
Grain-Size	188,500	.120	191,400	.142	.048	50-57	262	
8	234,000	.179	234,500	.179	.075	50-57	223	

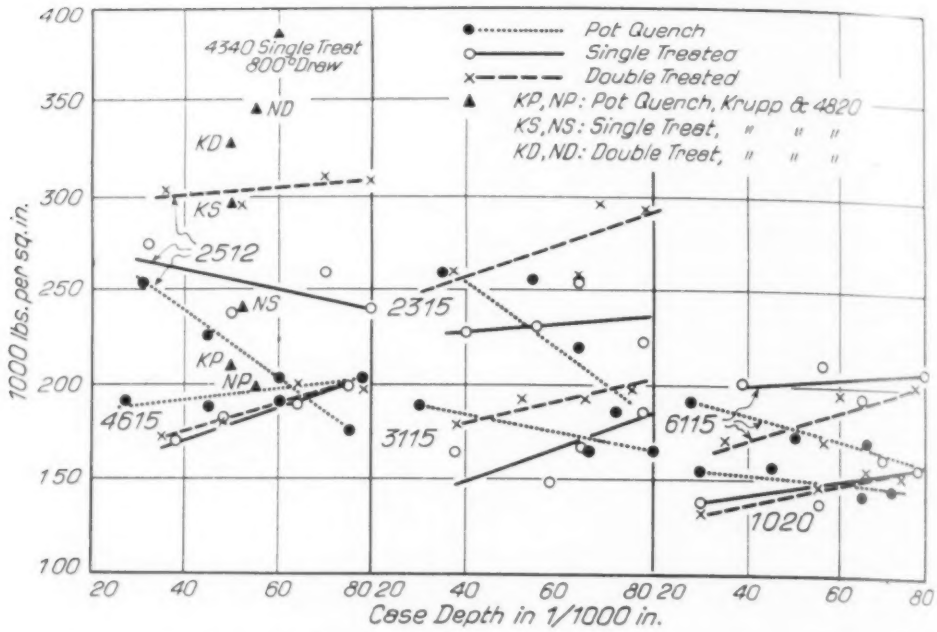

Fig. 16—Variation of Case Strength with Case Depth. Fine-Grained Steels—High Carbon Case Heat Treated as Shown.

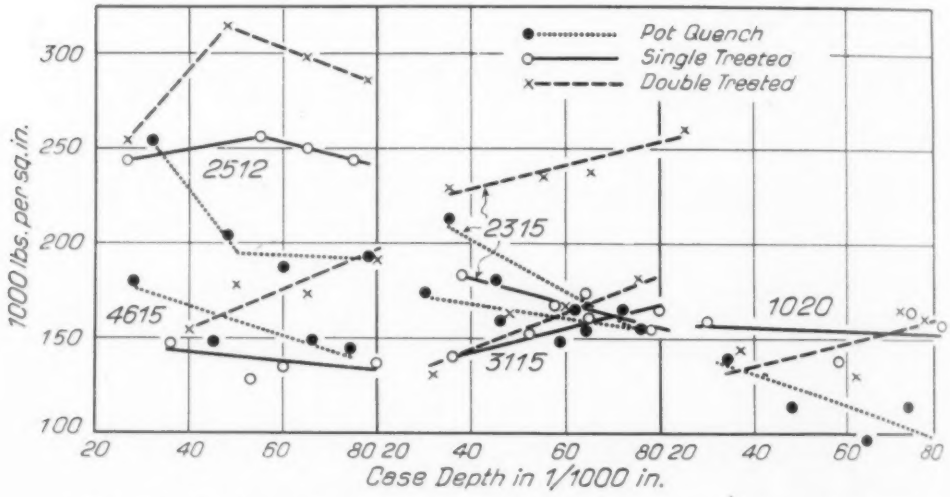

Fig. 17—Variation of Case Strength with Case Depth. Coarse-Grained Steels—High Carbon Case Heat Treated as Shown.

all of the steels, when double treated, increase in load carrying capacity with increase of case depth. This failure of the deeper case to increase load carrying capacity is probably due to internal stresses set up in hardening between the case and core, between the different zones of the case, or both.

Results of some tests tending to support this belief will be given

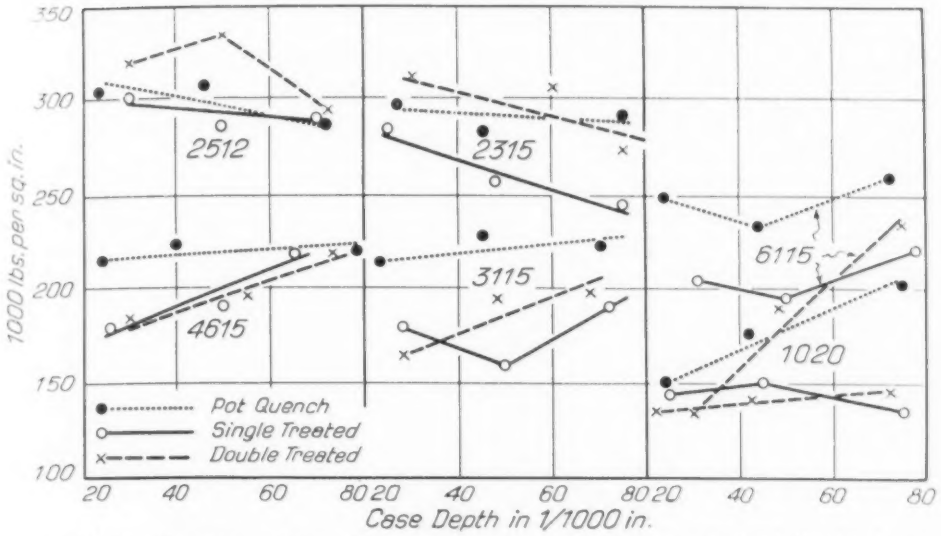

Fig. 18—Variation of Case Strength with Case Depth. Fine-Grained Steels—Low Carbon Case Heat Treated as Shown.

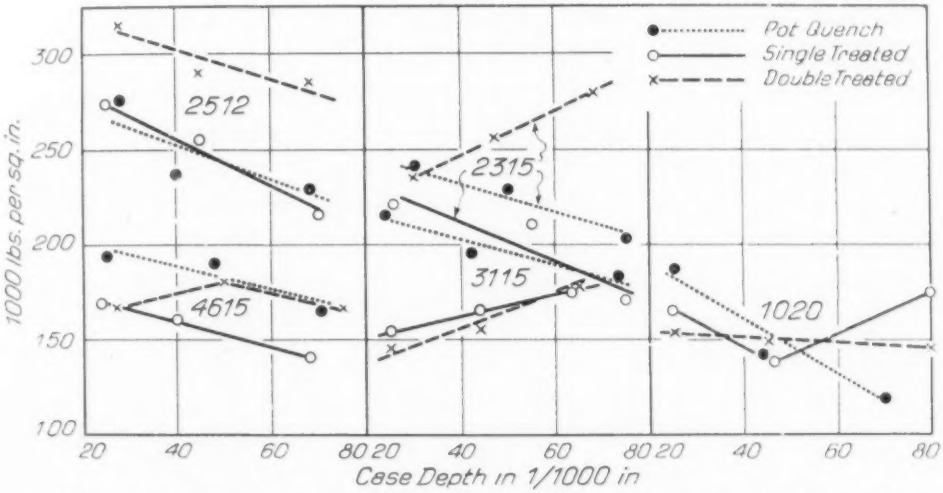

Fig. 19—Variation of Case Strength With Case Depth. Coarse-Grained Steels—Low Carbon Case Heat Treated as Shown.

later. It will be seen from Tables XI to XVI and Figs. 16-19, that double treatments, especially of the higher alloy steels and the coarse-grained steels, may very greatly increase the strength as compared to a pot quench or single reheat. The austenitic layer in the deeper cases, especially of the higher alloy steels when pot-quenched, may set up internal stresses between the zones of the case itself.

The Brinell hardness of the core does not appear to be an important factor in the strength of a case hardened part. There will be noticed a considerable variation in strength with the same

core hardness and examples may be picked out between types of steel where more than double the strength is obtained with a lower core hardness. The relationship between the case and the core and the structure of the core, that is, its uniformity and grain-size, are of more importance than hardness. The case depth values given in Tables XI to XVI were obtained by polishing and etching the broken ends of the hardened test pieces. Attempts were made to measure case depth by an ordinary eyepiece and with a Filar micrometer, but the most satisfactory results were finally obtained by the use of a Brinell glass. The fully hardened appearing part of the case was measured but this presented considerable difficulty, especially with the alloy steels where the line of demarcation between case and core is not distinct. The values are at least sufficiently close to indicate the trend.

In general it may be said that the strength of case hardened parts tends to follow the strength of the uncarburized steel. If not given the proper case depth and heat treatment, however, the advantages of the higher strength may be lost. For example, such strong steels as S.A.E. 2512 or Krupp, if deeply cased and pot quenched, may be but little stronger than case hardened S.A.E. 1020, but if double treated may be more than twice as strong. Greater care is apparently necessary in selecting the heat treatment for the high alloy steels since their strength in some instances was almost doubled by a change in the heat treatment alone, while plain carbon steel, especially when fine-grained, was little influenced as to static strength by either case depth or heat treatment. The coarse-grained steels are somewhat more erratic in behavior than the fine-grained steels and the latter are noticeably stronger than the corresponding coarse-grained steels. For the higher alloy steels, especially in the deeper cases, pot quenching gives the lowest results and such treatments should be avoided where great strength is desired, double treatments being by far the best. Case depth and heat treatment have less influence on the lower alloy steels but for deeper cases, pot-quenching gives lower results, with the one exception of fine-grained 4615 steel which appears to have ideal characteristics for that treatment. The coarse-grained 4615 behaves differently. Double treatments produce greatest strength in most of the steels with deep cases, and this tendency is more pronounced in the coarse-grained than in the fine-grained heats.

The lower carbon cases appear to be less influenced by case

Table XVII

Number of Cracks Occurring in the Case of Bend Test Specimens When Broken

Steel and Grain-Size	Number of Cracks in Case					
	High Carbon Case			Low Carbon Case		
	Pot Quench	Single Treat	Double Treat	Pot Quench	Single Treat	Double Treat
1020 Grain-Size 2–7	7	1	4 to 6	5 to 6	3 to 4	4 to 5
	3 to 6	1	3–4	3–4	2–3	3–4
	3–4	1	3	.....	.....	.....
	2–6	1	1–3	1	1	4
1020 Grain-Size 2–3	5–7	1–2	1–4	15–20	1–2	3
	4–7	1	2–7	10	1	4–5
	3–4	1	1–2	.....	.....	.....
	3–5	1	1–4	7–10	1	1–2
2315 Grain-Size 7	7	1–3	1	3–4	1	1
	1	1–2	1	1	1	1
	1	1	1	.....	.....	.....
	1	1	1	1	1	1
2315 Grain-Size 2–3	12	4–7	2–3	9	1–3	2
	8–10	2–6	2	6–7	1–3	1–2
	6	3	1	.....	.....	.....
	3–5	2–4	1	1–2	1	1
3115 Grain-Size 2–7	3–4	4–6	2–3	5–6	3–4	3–4
	3–4	4–5	1–2	1–2	3–4	1
	1	3–5	1	.....	.....	.....
	1	3–4	1	1	1–3	1–2
3115 Grain-Size 2–6	8–10	7	3–5	5–7	5–6	4–5
	8	5–8	2–3	5–6	5	4–5
	6	3–4	2–3	.....	.....	.....
	5	4	1	3	3–4	2
4615 Grain-Size 7–8	6–8	6–8	4–5	8–9	6	3
	7–8	3–4	1–3	5–6	3	2
	3–5	5	1–4	.....	.....	.....
	1–3	3–4	1–3	1	3	1–2
4615 Grain-Size 3	10	13	3–6	11–15	6–8	4
	8	8–10	3	8–10	7–8	4
	7–10	6–8	1–4	.....	.....	.....
	7	4–6	1–2	5–7	5	2–3
2512 Grain-Size 6–7	12	3	1	5	1	1
	10	1	1	1–3	1	1
	4–5	1	1	.....	.....	.....
	2–3	1	1	1	1	1
2512 Grain-Size 5	8–9	3–4	1–4	7–11	2	2
	8–10	1–2	1	2–3	1	1
	6–8	1	1	.....	.....	.....
	1–5	1	1	1	1	1
6115 Grain-Size 8	12	5–7	2–3	1	3–7	1–2
	12	1–2	2–3	1–3	1–2	1–2
	3–8	1–4	1–2	.....	.....	.....
	3–8	1–3	1	1	2	1

depth and heat treatment than do the corresponding high carbon cases. There is some difference in the results, especially the improvement of the position of pot-quenched samples perhaps due to the lower carburizing temperature, but in general they follow the same trends as the high carbon cases.

S.A.E. 2512, Krupp and 4820 steels show the highest load-carrying capacities when double treated. Next in order is S.A.E. 2315. The next are S.A.E. 6115, 4615, and 3115 in that order with not a great deal of difference between them. S.A.E. 1020 is weaker than the alloy steels.

Table XVIII

Bend Test Results on Square and Round Bars

Fine-Grained Steel Carburized on all Surfaces and Pot-Quenched

Steel	Case Strength at First Crack Pounds per Square Inch	Deflection at 1st Crack Inch	Maximum Case Strength Pounds per Square Inch	Deflection at Max. Load Inch	Case Depth Inch	Rockwell of Case	Core Brinell	Shape of Cross Section
4615	196,700	.122	217,000	.186	.032	60-62	293	Square
4615	225,200	.146	230,200	.159	.052	60	286	Square
4615	226,500	.135	226,500	.135	.072	56-59	277	Square
2512	277,500	.200	284,800	.228	.030	53-58	402	Square
2512	267,600	.181	267,600	.182	.046	52-56	402	Square
2512	215,000	.138	215,000	.139	.080	48-54	402	Square
4615	217,900	.119	266,800	.292	.045	61-62	277	Round
4615	204,700	.103	256,400	.252	.055	60-61	286	Round
4615	221,200	.113	226,800	.173	.084	57-59	277	Round
2512	272,500	.145	298,000	.180	.042	54-57	415	Round
2512	271,800	.142	270,900	.148	.055	52-55	415	Round
2512	249,100	.130	252,400	.130	.080	47-50	415	Round

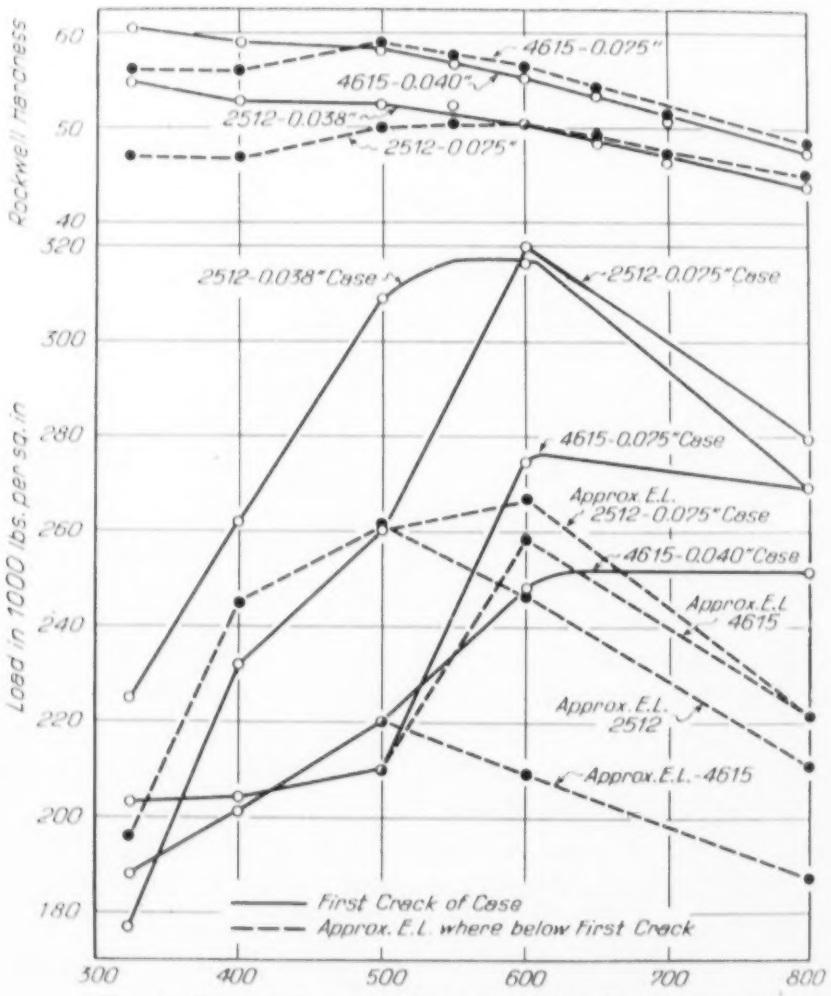

Fig. 20—Case Hardness and Strength of Case Hardened Bars.

Additional Bend Tests

A few bars of the same dimensions as previously given were carburized on all four sides and quenched from the box. Round bars 0.875-inch in diameter were carburized all over and given the same treatment. The results are given in Table XVIII. They follow the trend of previous results, indicating that neither the case on all sides nor the shape of the test specimen had any great influence on the results. The fractures also were similar in appearance to those previously listed.

Bars of fine-grained S.A.E. 4615 and 2512 with both shallow and deep cases were pot-quenched and drawn at temperatures of 400, 500, 600 and 800 degrees Fahr. The purpose of these treatments was to relieve internal stresses and leave only the effect of the depth of case. Table XIX and Fig. 20 give the results, which do show that strength does increase with case depth and also with moderate drawing temperatures as internal stresses are relieved. As the drawing temperature was raised, the first crack did not take place at the elastic limit, so that where the latter is different it has been drawn in as a broken line. The expected increase[2] in hardness from decomposition of austenite at higher drawing temperatures has taken place and the maximum of the elastic limit roughly coincides with the maximum Rockwell hardness. At drawing temperatures of 500 and 600 degrees Fahr. the fractures of the case appeared coarser than at either 400 or 800 degrees Fahr., and the cores produced mostly fine crystalline instead of silky fractures.

Results From Impact Tests

The heat treatment of impact specimens was identical to that of bars of the same analysis, an electric furnace being used. The scale from all the specimens was removed by scraping and polishing with emery cloth by hand. Considerable difference was noticed among the various steels as to the ease with which scale could be removed. This is of importance in cleaning such parts as gears, where wire brushing is a common method. The samples quenched direct from the carburizing temperature were, of course, all equally easy to clean as only a thin film of carbon and oil was present. Of the reheated samples, the S.A.E. 2512 samples were much the hardest

[2] H. W. McQuaid and O. W. McMullan, "Relation of Structure to Surface Hardness of a Case Hardened Steel," Transactions, American Society for Steel Treating, Vol. 18, 1930.

Table XIX

Effect of Drawing Temperature on Case Strength

From Bend Tests on Pot-Quenched Fine-Grained 4615 and 2512

Steel and Grain-Size	Drawing Temp. Degrees Fahr.	Case Strength at First Crack Pounds per Square Inch	Deflection 1st Crack Inch	Maximum Case Strength Pounds per Square Inch	Deflection at Max. Load Inch	Case Depth Inch	Rockwell of Case	Core Brinell
4615	400	201,800	.125	237,400	.363	.040	59	293
Grain 7-8	400	204,700	.118	217,600	.338	.075	56	302
2512	400	261,900	.170	311,000	.425	.038	53	401
Grain 6-7	400	232,100	.136	241,300	.167	.075	47	415
4615	500	220,500	.116	224,000	.121	.040	58.5	321
Grain 7-8	500	197,700	.093	204,700	.107	.075	59	311
2512	500	308,800	.230	308,800	.230	.038	52.5	401
Grain 6-7	500	260,400	.142	260,400	.142	.075	50	415
4615	600	248,900	.188	248,900	.188	.040	55.5	262
Grain 7-8	600	274,600	.157	274,600	.157	.075	56.5	262
2512	600	*	½ +	317,100	½ +	.038	50.5	401
Grain 6-7	600	320,100	.255	320,100	.255	.075	50.5	388
4615	800	*	½ +	247,300	½ +	.040	47.5	262
Grain 7-8	800	272,100	.314	272,100	.314	.075	48.5	265
2512	800	*	½ +	271,100	½ +	.038	44	331
Grain 6-7	800	280,500	.356	280,500	.356	.075	45	321

*Tore apart—no definite crack.

to clean, except for Krupp which was about the same. S.A.E. 2315 was next hardest and the 4340 the easiest. There was not much to choose among the others. The conclusion is that nickel increases the adherence of scale and makes cleaning more difficult.

Different individuals do not always agree as to the file hardness of a specimen. The size and shape of the specimen, the condition of the file, and amount of pressure have an influence. A broad flat surface may seemingly be harder to the file than a smaller size or cylindrical surface because of the greater unit pressure possible. It is difficult to get any surface that cannot be marked with the corner of a file. The impact specimens were used as a comparison for file hardness, since they were small and offered a better surface for attack as mentioned above, as well as being less influenced by furnace conditions because of the shorter time in the furnace. The tests given were quite severe and specimens were not considered file hard if there was any doubt. The samples quenched direct from the carburizing box were the hardest and in general were file hard, except for a slight skin on the surface. The S.A.E. 1020 was file hard except for a few soft spots. The S.A.E. 2512 was softest and the 2315 was softer than the 4615 and 3115, which were about equal. S.A.E. 6115 is unusual in that when pot-quenched, it has a fairly deep skin which is readily filed but it is very hard underneath this skin. Unless the surface were removed by grinding or lapping, a

Table XX

Impact Values on Case Hardened Specimens

Pot Quenched Samples—High Carbon Case—Drawn 1 Hour at 325 Degrees Fahr.

Steel and Grain-Size	Ft.-Lbs. Impact Notched	Ft.-Lbs. Impact Unnotched	Case Depth Inch	Rockwell of Case	Hardness to File
1020	5½	11	.030	63-64	File hard.
Grain-Size	4½	9½	.028	64-66	
2-7	4	8	.060	63-65	
1020	4	17	.028	62-65	File hard.
Grain-Size	4	14	.030	65	
2-3	3½	8	.060	63-64	
2315	12	38	.028	58-60	Soft skin. Nearly hard underneath.
Grain-Size	7½	23½	.042	51-54	
7	6	20	.065	51-57	
2315	14	48	.028	60-61	Soft skin. Nearly hard underneath.
Grain-Size	10	25	.042	53-57	
2-3	8½	16½	.058	53-56	
3115	7	20	.027	62-63	Practically file hard. Very thin skin. Hard underneath.
Grain-Size	5½	12	.040	56-58	
2-7	4	8	.060	56-60	
3115	11½	35	.028	61-63	Practically file hard. Very thin skin. Hard underneath.
Grain-Size	8	16	.040	56-60	
2-6	6	10	.060	56-60	
4615	14	27	.030	62-63	Practically file hard. Very thin skin. Hard underneath.
Grain-Size	7½	17	.040	56-59	
7-8	5½	13	.060	57-60	
4615	7½	23	.030	61-62	Practically file hard. Very thin skin. Hard underneath.
Grain-Size	7	15	.044	57	
3	5	9	.060	56-58	
2512	11	33	.030	53-56	Not file hard. Deep skin. Can be filed with difficulty underneath.
Grain-Size	8	26	.044	47-53	
6-7	6½	21	.060	47-51	
2512	9	26	.030	56-58	Not file hard. Deep skin. Can be filed with difficulty underneath.
Grain-Size	6	19	.042	51-54	
5	5½	15	.060	48-52	
6115	9	20	.026	63	Surface quite soft. Hard underneath.
Grain-Size	5	12	.040	55-57	
8	4	8½	.060	51-56	
Krupp Gr.-7	8½	33	.044	45-53	Hard except for slight skin.
4820 Gr.-7	9	28½	.045	57-60	Very thin soft skin.
4340	2	11	.050	56-57	400° Draw. Rather deep soft skin.
Grain-Size 6-7	2½	13½	.050	51	800° Draw. Not hard.
Cr-Ni-Mo. Grain-4	8	18	.048	50-53	Hard except for slight skin.
Cr-Ni-Mo. Grain-4	5½	14½	.048	49-51	Hard except for slight skin.

pot quench would not be recommended for this steel if file hardness is desired. None of the thin cases on the low carbon treatments were file hard. While steel of eutectoid composition or lower will give high Rockwell readings when hardened, it is necessary to have a high carbon content to obtain file hardness. Partial decarburization which occurs from reheating in an ordinary hardening furnace, is usually sufficient to produce a thin layer which may be caught with a file. Most of the single treated samples were file hard or nearly so underneath the decarburized layer except the 2315 and 2512, which could be filed deeper but not readily. Under the decarburized layer,

Table XXI

Impact Values on Case Hardened Specimens

Single Treated Samples—High Carbon Case—Drawn 1 Hour at 325 Degrees Fahr.

Steel and Grain-Size	Ft.-Lbs. Impact— Notched	Unnotched	Case Depth Inch	Rockwell of Case	Hardness to File
1020	4	7	.028	60-63	Nearly file hard.
Grain-Size	2	6	.035	62-64	
2-7	2	6	.053	61-64	
1020	2½	5	.030	62-64	Nearly file hard.
Grain-Size	2	4	.040	60-64	
2-3	1½	4	.054	58-64	
2315	9½	51½	.036	57-59	Not hard. Soft surface.
Grain-Size	7	40½	.050	55-58	
7	5½	22½	.070	56-60	
2315	11	26	.035	59	Not hard. Soft surface.
Grain-Size	9½	19	.052	58-59	
2-3	6½	11	.075	57-59	
3115	3½	8	.038	62-63	Nearly file hard.
Grain-Size	2½	8	.052	61-62	
2-7	2	5½	.070	61-63	
3115	4	7	.033	60-63	Nearly file hard.
Grain-Size	2½	5½	.047	62-63	
2-6	2½	5½	.065	60-63	
4615	16½	29½	.042	61	Nearly file hard.
Grain-Size	8	16	.050	59-62	
7-8	4	7	.078	61-63	
4615	5	14	.037	60-61	Nearly file hard.
Grain-Size	3½	6½	.048	60-62	
3	3	5	.072	60-62	
2512	8½	47½	.032	56-57	Not file hard.
Grain-Size	6½	25	.045	55-57	
6-7	5	14	.060	56-57	
2512	5	24	.032	56-58	Not file hard.
Grain-Size	4	14	.046	56-57	
5	3½	7½	.068	55-58	
6115	8½	22	.035	60-64	Hard except for soft skin.
Grain-Size	4	11	.050	60-63	
8	3	9½	.072	61-63	
Krupp Gr.-7	7½	20	.045	60-61	Rather deep soft skin.
4340	2	7	.050	62	400° Draw. Rather deep soft skin.
Grain-Size-6-7	2	17	.050	52-54	800° Draw. Not hard.
4820 Gr.-7	6½	25	.050	59-60	Nearly file hard.
Cr-Ni-Mo. Grain-4	8	14	.052	62-63	Rather deep soft skin. Hard underneath.
Cr-Ni-Mo. Grain-4	4½	9	.052	62-63	Rather deep soft skin. Hard underneath.

S.A.E. 6115 was hardest but not much more so than 3115 and 4615 in the order named, and 2512 was softest. The thin cases of low carbon content all could be filed. The double treated samples were similar except that the partially decarburized layer was deeper. It may, therefore, be said that high carbon in a case is necessary for absolute file hardness and either a low initial carbon content or partial decarburization from reheating not sufficient to be noticeable on the Rockwell, may produce a surface soft to the file. This condition was more noticeable on the larger samples which were held in the furnace longer. The effect of nickel is to decrease file hardness, but it must

be kept in mind that all samples were drawn at the same temperature, 325 degrees Fahr., and case hardened nickel steels soften at low drawing temperatures. However, in some later tests no noticeable difference was found in file hardness between 2512 as pot-quenched and after tempering at 325 degrees Fahr.

The impact specimens, which were standard 10-millimeter square Izod bars made long enough for each to give two notched and two unnotched tests, were broken so that one blow of each specimen was made at right angles to the other. The results are given in Tables XX to XXV. As to be expected, impact values decreased with increase in case depth in both the notched and unnotched condition. There was found to be no uniform relationship between notched and unnotched values, in some cases the notched impact being nearly equal

Table XXII

Impact Values on Case Hardened Specimens

Double Treated Samples—High Carbon Case—Drawn 1 Hour at 325 Degrees Fahr.

Steel and Grain-Size	Ft.-Lbs. Impact Notched	Ft.-Lbs. Impact Unnotched	Case Depth Inch	Rockwell of Case	Hardness to File
1020	10	11	.028	56-62	Not file hard. Decarburized surface.
Grain-Size	5½	11	.035	56-63	
2-7	4	11	.050	58-63	
1020	5	8½	.028	56-62	Not file hard. Decarburized surface.
Grain-Size	3½	7	.040	56-65	
2-3	3	6½	.056	59-63	
2315	13	85	.038	57-60	Not file hard.
Grain-Size	9	66	.050	59-60	
7	7	58	.064	58-60	
2315	14½	49½	.040	58-60	Not file hard.
Grain-Size	13½	37	.050	58-60	
2-3	9	17	.065	59-61	
3115	16	25	.028	61-62	Nearly file hard.
Grain-Size	10	20	.045	58-62	
2-7	8	14	.061	58-62	
3115	16	30	.030	60-62	Nearly file hard.
Grain-Size	12½	19	.040	60-63	
2-6	9	10½	.060	60-63	
4615	19	29½	.032	59-62	Nearly file hard. A little softer
Grain-Size	15½	30	.045	60-63	than 3115.
7-8	11	18½	.060	60-63	
4615	21½	39	.028	60-62	Nearly file hard. A little softer
Grain-Size	16	28	.035	61-63	than 3115.
3	9	16	.060	60-62	
2512	12	94	.038	56-58	Not file hard.
Grain-Size	7	54	.050	57-58	
6-7	6½	34	.065	57-59	
2512	6	48½	.030	56-58	Not file hard.
Grain-Size	5	34	.047	57-60	
5	4	19	.060	57-59	
6115	15	37	.030	54-62	Hard except for decarburized surface.
Grain-Size	9	15	.045	61-63	
8	5	10	.064	62-64	
Krupp Gr.-7	10	34	.045	61-62	Not file hard.
4820 Gr.-7	8½	49	.045	62	Nearly file hard.
4340	1½	7	.050	63	400° Draw. Slight soft skin.
Grain-Size	2½	13	.050	53	800° Draw. Not hard.
6-7					

to the unnotched on the same specimen, while in other cases the unnotched value was many times the notched in both shallow and deep cases. The tables show that the nickel steels may lose much of their superior shock resistance if notches are present in the case and should be avoided in these steels more than others, especially when single or double treated. There is not much difference among the other steels, but 1020 is slightly less affected by notches, the coarse-grained 1020 being affected more than the fine.

Almost without exception, single treatments give lower impacts than pot-quenched or double treated samples. The single treated S.A.E. 6115 is better than the pot-quenched. Since the impact is lower, the strength of the bars in the bend test is lower, and they develop a soft skin, a direct quench from the carburizing temperature is not a satisfactory treatment for S.A.E. 6115 steel. Those steels such as 4615 and 6115 which were single treated by quenching from above the critical range of their core are less adversely affected than those quenched from within the range such as S.A.E. 1020 and 3115. This brittleness resulting from quenching within the critical range has been reported by the author in a previous paper.[3]

It generally has been found that a coarse-grained steel will have its properties more improved by double treating than will a fine-grained steel as compared to a single quench. A comparison of the impacts for these treatments on the coarse and fine 1020 and 4615 will demonstrate this. All the steels, however, tend to show higher impact values when double treated. There is no such thing as a double treatment in which the second treatment affects only the case and not the core, since the Ac_1 on both are practically the same. The advantage of the double treatment is that instead of leaving the core of such steels as S.A.E. 1020 and 3115 as a mixed structure of coarse grains of ferrite and martensite as is produced by a single reheat at a low temperature, a fine-grained uniform structure is produced. These structures are due in the one case to a starting product of coarse ferrite and pearlite, and in the other to a case structure refined by the first quench.

The generally accepted statement that fine-grained steels have higher impact values than coarse-grained steels has not been entirely confirmed. One prominent exception is that of the two heats of S.A.E. 2315, which were very good examples of a fine and coarse

[3] O. W. McMullan, "Effect of Quenching Temperature Change on the Properties of Quenched Steel," TRANSACTIONS, American Society for Steel Treating, Vol. XIV, October 1928.

Table XXIII

Impact Values on Case Hardened Specimens

Pot Quenched Samples—Low Carbon Case—Drawn 1 Hour at 325 Degrees Fahr.

Steel and Grain Size	Ft.-Lbs. Impact Notched	Ft.-Lbs. Impact Unnotched	Case Depth Inch	Rockwell of Case	Hardness to File
1020	7	12	.018	51-60	Not hard.
Grain Size	7½	9	.040	64-65	File hard.
2-7	4	9	.060	56-64	Spotted.
1020	4½	33	.020	60-63	Hard.
Grain Size	4	13	.040	63-65	Hard.
2-3	3	4½	.060	62-64	Spotted.
2315	14	52	.020	57-61	Soft skin. Nearly hard underneath.
Grain Size	8½	28	.040	60-61	
7	4	15	.070	57-60	
2315	15½	57	.022	56-60	Soft skin. Nearly hard underneath.
Grain Size	10½	25½	.040	61	
2-3	7	11½	.070	60-62	
3115	9½	23	.018	56-63	Hard except thin skin.
Grain Size	5½	12	.045	63	
2-7	3½	9	.070	61-63	
3115	20	48½	.020	55-58	Hard except thin skin.
Grain Size	10½	14	.040	63-64	
2-6	3½	7	.075	61-63	
4615	23	64	.020	55-61	Hard except thin skin.
Grain Size	12	20	.040	63-64	
7-8	4½	13½	.070	63	
4615	10	30	.022	57-60	Hard except thin skin.
Grain Size	5	9½	.040	62-63	
3	2½	5	.070	61-62	
2512	11½	45	.020	56-58	Soft skin. Can be filed with difficulty underneath.
Grain Size	9½	37	.038	59.5	
6-7	5½	20½	.070	56	
2512	8	30	.020	58	Soft skin. Can be filed with difficulty underneath.
Grain Size	6	19½	.040	59-60	
5	4½	11½	.070	57-58	
6115	5	24	.017	55-64	Hard except for thin skin.
Grain Size	3	15	.042	64-65	Not as deep as on higher carbon cases.
8	2	4	.070	60-64	

grain. No matter what the treatment given it, the coarse-grained steel had much higher impact values than did the fine for the same case depth. This may have been due partially to the lower hardening power of the coarse-grained steel because of its lower carbon and manganese content, but there was much less difference in the core hardness of the impact specimens than in the bend tests, the maximum difference being only 3 to 4 points Rockwell C for any one treatment. Neither can it be said that types of steel that are naturally fine-grained have higher impact values than those more apt to be coarse. Although S.A.E. 6115 is naturally fine-grained and in these experiments was the finest grained steel tested, yet it developed some of the lowest impact values and did not live up to its reputation of being highly shock resistant.

The carbon content of the case appears to have but little effect on the impact values, except perhaps in the unnotched condition of

some steels when single or double treated. The depth of case has the greatest influence on impact results. As a better means of more readily comparing the different steels, the average values of impact results on any one steel and grain-size, irrespective of case depth or carbon content, have been listed in Table XXVI and plotted in Fig. 21. The results may not be exactly representative for a particular case depth but represent the general trend.

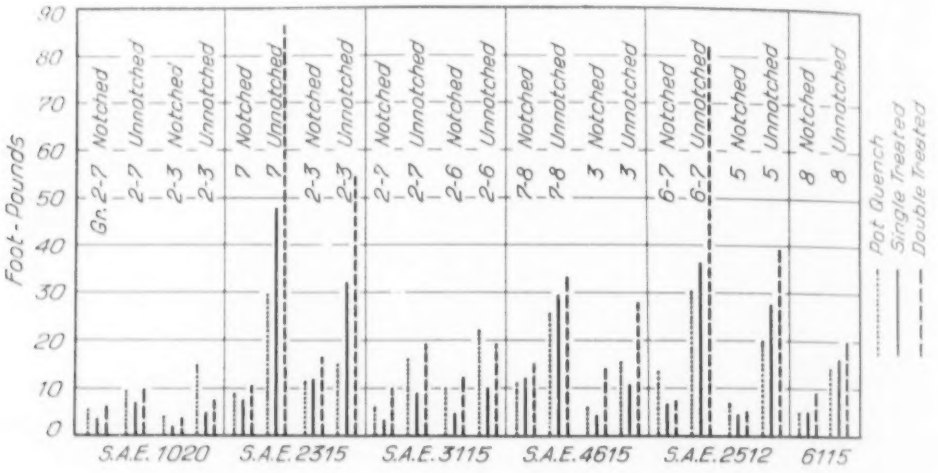

Fig. 21—Average of All Impact Results on Case Hardened Steels.

The fractures of the pot-quenched impact specimens were very similar to those on the corresponding bars. On the single and double treated samples of plain carbon steels and those of lower alloy content there was some tendency for the impact specimens to have a fine crystalline fracture in the core, whereas the bars were silky. This was probably due to the greater effect of the quench on the small sized specimens and also to the fact that they were carburized on all four sides. It may, therefore, be said that the rate of breaking of a case hardened specimen has little or no influence on the appearance of the fracture. There was no difference in appearance of the fracture at the notch and where not notched. The notches were made before carburizing and heat treatment. It has been found that a notch made in the case after treatment frequently does have an effect on the appearance of the fracture.

Distortion Tests

The test specimens, of the design shown in Fig. 14, were all quenched in the same manner with the slotted end down. The parts

Table XXIV

Impact Values on Case Hardened Specimens

Single Treated Samples—Low Carbon Case—Drawn 1 Hour 325 Degrees F.

Steel and Grain-Size	Ft.-Lbs. Impact Notched	Ft.-Lbs. Impact Unnotched	Case Depth Inch	Rockwell of Case	Hardness to File
1020	6	9½	.015	55-60	Not hard.
Grain-Size	3½	7	.038	63	Nearly hard.
2-7	2½	8	.052	63-64	Nearly hard.
1020	2½	6	.021	61-62	Nearly hard.
Grain-Size	2½	5½	.040	63	
2-3	1½	5	.060	63-64	
2315	11½	74	.024	57-59	Not file hard.
Grain-Size	7	40	.045	59-61	
7	3½	56½	.070	57-59	
2315	31	97	.025	56-57	Not file hard.
Grain-Size	9	23	.050	59-61	
2-3	5	15	.080	58-59	
3115	5½	13	.023	59-61	Nearly file hard.
Grain-Size	3	10	.044	62-64	
2-7	2	6½	.080	61-64	
3115	11½	26½	.020	58-61	Nearly file hard.
Grain-Size	4½	10	.045	61-63	
2-6	2	7	.080	60-62	
4615	25	74	.024	58-60	A little softer than 3115.
Grain-Size	13	33	.045	61-62	
7-8	4	17	.080	60-62	
4615	5½	17	.030	58-61	A little softer than 3115.
Grain-Size	4	13	.050	60-61	
3	2	7	.080	59-61	
2512	9½	50	.024	57-58	Not file hard.
Grain-Size	6	42	.046	58-59	
6-7	4	37	.075	56-58	
2512	6½	59	.025	55-58	Not file hard.
Grain-Size	5	35½	.048	57-59	
5	3	26	.080	56-57	
6115	9	24½	.026	58-61	Not file hard.
Grain-Size	3½	19	.052	59-63	Nearly file hard.
8	2	9	.080	59-63	Nearly file hard.

were carefully cleaned of scale before measurements were made. All specimens were carburized in the same containers as the corresponding impact specimens and since they were of somewhere near the same size as those specimens, they should have approximately the same depth of case as given in Tables XX to XXV. No sections were cut from the samples for microscopic examinations.

The distortion results obtained are plotted in Figs. 22 to 24. The lines extending to the left of the zero base line represent the amount of closing in at the end of the prongs from that of the original untreated sample and spreading out is measured to the right. The vertical column of figures adjacent to each group of curves denote in ten thousandths of an inch, the maximum difference in change of dimension between the two pairs of corners (a to b and c to d of Fig. 14) and therefore represent the maximum amount of twisting of the prongs with respect to each other but not with

the body of the sample. In the pot-quenched samples the horizontal lines show the average change across corners after final treatment. In the single and double treated samples the horizontal lines denote the change after carburizing and cooling in the box. The slanting

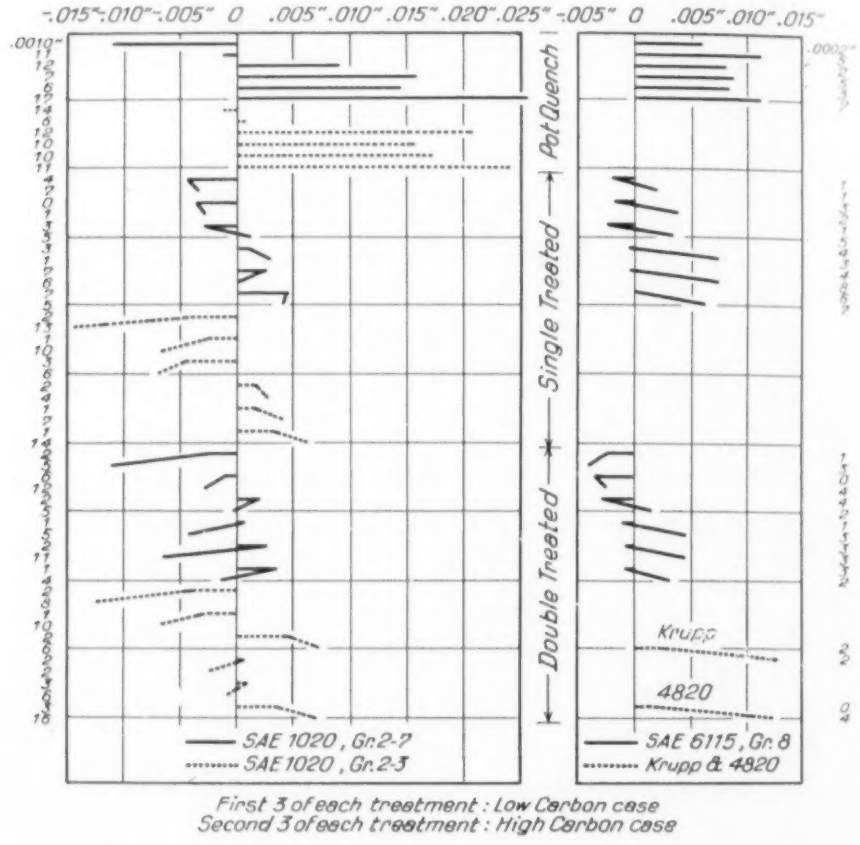

Fig. 22—Distortion of Carburized and Hardened Test Blocks.

lines show the further change from that point after final heat treatment in both the single and double treated samples, the lower end of the line being the final position.

There is not a great deal of difference in the behaviors of several of the alloy steels but perhaps the specimens were not properly designed to bring out the differences. One of the chief things observed was the remarkable regularity of the changes which occurred even in the water-quenched S.A.E. 1020 when conditions were kept identical. Much of the trouble arising from variation in distortion may, therefore, be due not to the steel itself but to variations in the carburizing and hardening procedure. There was some difference

in behavior between the fine-grained and coarse-gained 1020 but both followed the same general trend. But little influence of grain-size was observable in the alloy steels. Especially when single or double treated, case depth is the greatest factor in differences of distortion. This has been the experience also with production parts. Except on the plain carbon steel, the case depth has but little influence on distortion when the parts are pot-quenched. This is perhaps due

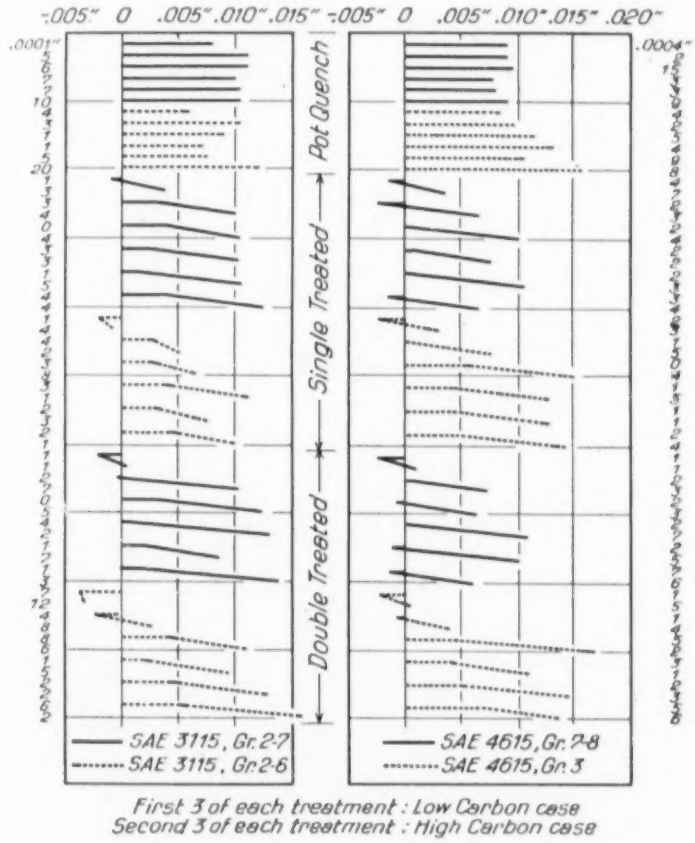
-.005" 0 .005" .010" .015" -.005" 0 .005" .010" .015" .020"
.0001"
.0004"
Pot Quench
Single Treated
Double Treated
SAE 3115, Gr. 2-7
SAE 3115, Gr. 2-6
SAE 4615, Gr. 2-8
SAE 4615, Gr. 3
First 3 of each treatment : Low Carbon case
Second 3 of each treatment : High Carbon case

Fig. 23—Distortion of Carburized and Hardened Test Blocks.

to the greater influence of the core when more fully hardened by the high quenching temperature. If distortion is to be kept uniform among a number of parts, it would appear that it is of most importance to keep the case uniform and of constant depth, and that this factor becomes of less importance if the parts are quenched direct from the carburizing box. In the figures three case depths are plotted for each treatment. The thinnest case is the top line in each group of three, the bottom the deepest.

Resistance of Cases to Crushing

Case hardened parts frequently are reported as having failed by crushing of the case into the soft core. An attempt was made by the use of hardened tool steel cylinders to approach the loading conditions obtained at the contact of teeth in gears. It was found to be difficult to get the load uniform along the cylinder and prevent

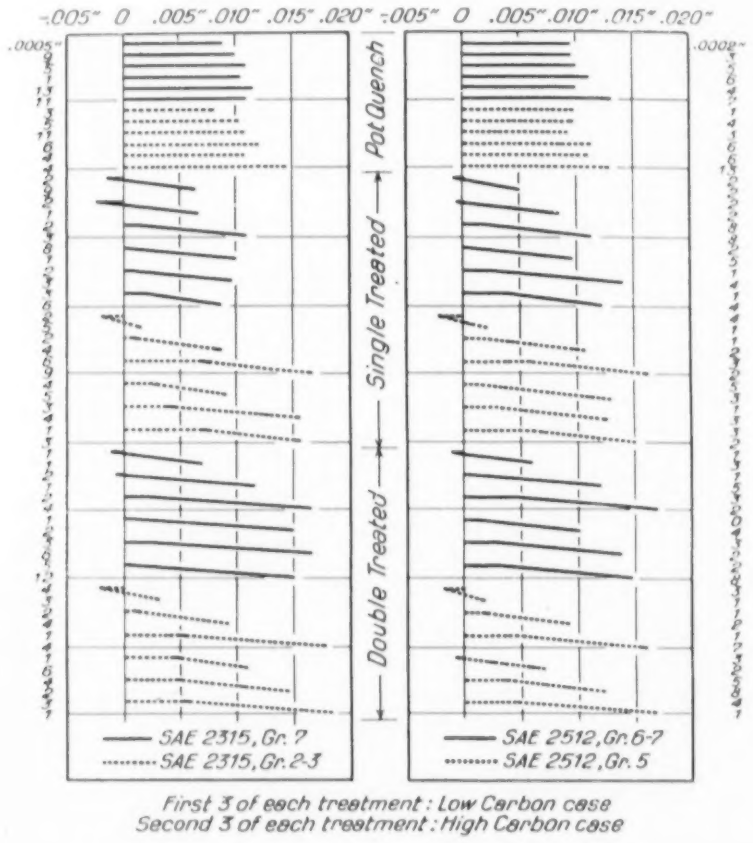
-.005" 0 .005" .010" .015" .020"
-.005" 0 .005" .010" .015" .020"
.0005"
.0002"
Pot Quench
Single Treated
Double Treated
SAE 2315, Gr. 7
SAE 2315, Gr. 2-3
SAE 2512, Gr. 6-7
SAE 2512, Gr. 5
First 3 of each treatment: Low Carbon case
Second 3 of each treatment: High Carbon case

Fig. 24—Distortion of Carburized and Hardened Test Blocks.

greater sinking at one end. Enough results were obtained by the use of pivoted blocks to show that results obtained were of the same order as those from the use of Brinell balls. Special alloy Brinell balls of high load carrying capacity were used and the impressions made in the undistorted ends of the ¾-inch square bars used for bend tests. Loads in steps of 5000 pounds were applied from 5000 to 25,000 pounds. Loads under 5000 pounds produced impressions too small to read with sufficient accuracy. The impressions made with the lower loads were all from new balls before any high loads

Table XXV

Impact Values on Case Hardened Specimens

Double Treated Samples—Low Carbon Case—Drawn 1 Hour 325° F.

Steel and Grain-Size	Ft. Lbs. Impact Notched	Ft. Lbs. Impact Unnotched	Case Depth Inch	Rockwell of Case	Hardness to File
1020	9	12	.013	53-59	Not file hard.
Grain-Size	7	7	.040	58-63	
2-7	5	9½	.051	55-63	
1020	7	10	.020	56-59	Not file hard.
Grain-Size	3½	7	.034	58-63	Nearly hard.
2-3	2½	6½	.053	57-63	Nearly hard.
2315	22	120	.024	55-57	Not file hard.
Grain-Size	7	86	.040	57-61	
7	5½	104	.080	58-59	
2315	44	120	.025	52-56	Not file hard.
Grain-Size	12½	57	.050	57-59	
2-3	5	42	.080	58-59	
3115	12	23½	.024	59-61	Nearly file hard.
Grain-Size	8½	18	.045	62-63	
2-7	5	13	.075	60-63	
3115	18	32	.020	51-60	Nearly file hard.
Grain-Size	11	17	.038	59-62	
2-6	5	8	.073	60-62	
4615	24	71	.020	51-58	A little softer than 3115.
Grain-Size	14	31	.040	60-63	
7-8	7	19	.072	58-63	
4615	22	46	.020	56-59	A little softer than 3115.
Grain-Size	13	23	.038	59-62	
3	4	14	.060	61-63	
2512	11½	106	.030	55-58	Not file hard.
Grain-Size	5½	110	.048	56-59	
6-7	4	96	.080	58-59	
2512	7	62½	.028	54-57	Not file hard.
Grain-Size	6	35	.045	58-59	
5	3½	37	.075	57-59	
6115	13½	26½	.020	54-59	Not file hard.
Grain-Size	8	19	.040	59-63	Nearly file hard.
8	4	13	.088	60-63	Nearly file hard.

were applied, since flat spots would affect the small impressions more than the larger. The special balls stood up very well even under high loads. The impressions were read by a Brinell microscope and converted to Brinell hardness numbers by the formula given by Herman A. Holz in Bulletin No. 27. The calculated readings for each load are given in Tables XXVII to XXXII. The load at which the first crack could be heard is given also and if it could not be heard, the lowest reading at which it was seen by the microscope when reading the impression, the latter figures being those of multiples of 5000 pounds in the Tables.

The use of the Brinell ball made possible uniformity of loading in all the tests and permitted results to be converted to an established hardness scale. Since the Brinell hardness number should be approximately the same on any one piece regardless of load, any great drop in hardness at the higher loads shows that the case is no longer

Table XXVI
Average of Impact Values Taken From Tables XX to XXV

Treatment	Steel	Fine-Grained		Coarse-Grained	
		Notched	Unnotched	Notched	Unnotched
Pot Quench	1020	5½	9½	4	15
	2315	8½	29½	11	31
	3115	6	16	10	22
	4615	11	25½	6	15
	2512	8½	30½	6½	20
	6115	4½	14		
Single	1020	3½	7	2	5
	2315	7½	47½	12	32
	3115	3	8½	4½	10
	4615	12	29½	4	10½
	2512	6½	36	4½	27½
	6115	5	16		
Double	1020	6½	10	4	7½
	2315	10½	86½	16½	54
	3115	10	19	12	19½
	4615	15	33	14	27½
	2512	7½	82	5	39
	6115	9	20		

capable of carrying the load. The thinnest cases may also give a uniform, but lower, hardness if the case is not deep enough to have much effect. This can be noticed by comparing hardness values on thin cases with those on the deeper cases of the same steel and treatment. With the higher alloy steels, especially when pot-quenched, the effect of the austenitic surface layer is to give lower readings from the lighter loads. Where the case depth was sufficient to withstand the load, the maximum Brinell hardnesses on the various steels and treatments corresponded to their relative positions in Rockwell hardness. With higher loads, expect on the deep cases where crushing did not occur, the Brinell hardness readings on the case depended upon the hardness of the core. The higher alloy steels such as S.A.E. 2512, therefore, do not require as deep a case to resist loads which may crush through into the core. Along with the drop in hardness, cracking and sinking of the case could be noticed around the impression. No sinking of the case outside of the impression itself could be observed at the lower loads.

The results with both the Brinell ball and the hardened cylinders lead to the conclusion that even thin cases practically never fail under ordinary conditions by the case crushing through into the core, particularly on such case hardened parts as gears where the area of contact between the surfaces is relatively small. Long before the loads become great enough to crush the case, the Brinelling action in the outer part of the case itself would destroy bearing surfaces and make the parts noisy and unfit for use.

Table XXVII

Resistance of Case to Crushing Under a Brinell Ball

Pot Quenched Samples—High Carbon Case

Steel and Grain Size	Case Depth Inch	Brinell Hardness from Applied Load of— 5,000 Pounds	10,000 Pounds	15,000 Pounds	20,000 Pounds	25,000 Pounds	Core Brinell	Load at First Crack Pounds
1020	.030	402	347	352	351		217	5,000
Grain-Size	.045	563	568	547	514		255	
2-7	.065	680	650	629	593		262	
	.072	680	650	611	578		255	
1020	.036	563	516	506	460		293	6,400
Grain-Size	.048	618	628	595	566		285	20,000
2-3	.064	648	672	647	623		277	
	.074	648	672	647	623		285	
2315	.033	494	516	506	493		402	
Grain-Size	.054	494	516	547	538		402	
7	.069	590	587	595	593		415	
	.072	538	587	595	593		430	
2315	.035	494	470	467	451		321	
Grain-Size	.045	538	550	520	538		331	
2-3	.062	563	587	579	566		321	
	.072	590	587	579	566		341	
3115	.030	516	484	467	460		302	15,000
Grain-Size	.048	538	550	533	514		302	20,000
2-7	.066	590	587	579	566		311	
	.080	563	628	647	605		321	
3115	.030	418	407	404	378		269	10,000
Grain-Size	.046	538	516	493	469		262–269	20,000
2-6	.064	590–618	587	547	538		285	20,000
	.076	590	587	579	551		277	
4615	.027	402	407	386	378		302	10,000
Grain-Size	.045	516	516	493	481		311	20,000
7-8	.060	590	550	520	493		269	
	.078	618	628	629	605		311	
4615	.028	436	407	404	396		277	5,000
Grain-Size	.045	563	550	520	481		277	20,000
3	.066	590	606	595	566		293	
	.074	590	606	579	566		269	
2512	.031	454	457	467	460	438	415	
Grain-Size	.045	494	516	520	514	525	402	
6-7	.060	494	532	563	538	525	415	
	.075	474	532	547	566	563	402	
2512	.032	474	470	467	469		388	
Grain-Size	.058	494	516	520	502		388	
5	.068	538	550	547	526		375	
	.085	516	550	547	538		388	
6115	.028	436	396	386	378		302	11,400
Grain-Size	.050	538	532	520	514		302	
8	.066	648	650	629	593		302	
	.080	590	628	629	605		293	
Krupp Gr.-7	.050	494	484	493	493		363	
4820 Gr.-7	.055	538	550	579	551		415	
4340	.054	538	587	611	605		514—400° Draw	
Grain 6-7	.060	474	484	493	493		430—800° Draw	
Cr-Ni-Mo Grain 4	.050	538	550	535	514		415	
Cr-Ni-Mo Grain 4	.050	563	587	579	578		495	

The writer offers the theory that on case hardened parts under compression at right angles to the surface, such as gear teeth, that cracking, pitting and spalling of the case is not due to the case crushing into the core unless the case is very thin, but rather that the failure is due to fatigue. When the case is relatively thin because

Table XXVIII
Resistance of Case to Crushing Under a Brinell Ball
Single Treated Samples—High Carbon Case

Steel and Grain-Size	Case Depth Inch	Brinell Hardness from Applied Load of— 5,000 Pounds	10,000 Pounds	15,000 Pounds	20,000 Pounds	25,000 Pounds	Core Brinell	Load at First Crack Pounds
1020	.030	436	386	352	324		179	10,600
Grain-Size	.055	516	484	404	378		187	9,800
2-7	.070	648	568	520	481		192	20,000
	.078	680	587	547	493		179	19,800
1020	.030	474	431	413	378		196	13,600
Grain-Size	.058	618	568	506	469		192	15,800
2-3	.075	648	628	547	502		202	20,000
	.082	680	532	563	526		187	19,000
2315	.040	474	484	493	502		340	
Grain-Size	.055	538	568	563	551		364	
7	.064	563	568	563	566		388	
	.078	590	606	611	578		364	
2315	.038	454	444	444	430		269	
Grain-Size	.058	538	516	520	493		269	
2-3	.064	563	516	506	481		248	
	.078	563	568	563	538		269	
3115	.038	494	470	467	430		277	12,300
Grain-Size	.058	563	550	520	493		302	15,000
2-7	.065	590	550	547	526		302	
	.078	618	606	595	566		269	20,000
3115	.036	474	431	424	387		235	11,400
Grain-Size	.052	538	501	467	451		235	15,800
2-6	.065	590	550	520	481		228	16,500
	.080	590	587	563	538		235	
4615	.038	418	386	361	351		255	9,100
Grain-Size	.048	494	418	395	378		229	12,400
7-8	.064	590	550	520	469		262	20,000
	.075	590	550	506	469		235	
4615	.036	494	444	424	405		262	10,000
Grain-Size	.053	538	516	493	460		241	10,000
3	.060	563	516	506	460		241	15,000
	.080	618	606	595	538		253	15,000
2512	.032	474	484	493	493		388	
Grain-Size	.050	538	532	533	538		388	
6-7	.070	563	587	579	566		388	
	.088	538	569	579	566		375	
2512	.028	474	470	467	469		415	
Grain-Size	.055	538	532	533	514		388	
5	.065	538	550	563	551		402	
	.075	563	568	579	566		364	
6115	.039	516	501	479	460		286	13,600
Grain-Size	.056	563	532	506	493		262	
8	.065	590	550	547	514		262	
	.080	618	628	595	566		262	
Krupp Gr.-7	.050	590	568	563	551		415	
4820 Gr.-7	.052	590	587	579	566		401	
4340	.054	618	628	647	656		514—400° Draw	
Grain 6-7	.060	494	501	506	502		430—800° Draw	
Cr-Ni-Mo Grain 4	.050	590	568	547	538		364	
Cr-Ni-Mo Grain 4	.050	618	628	629	623		477	

of the lower hardness and endurance limit of the core, failure starts in the core next to the case (See Fig. 15) and proceeds outward through the case rather than inward from the surface. The value of a harder core would be its higher endurance limit rather than its ability to prevent the case from being pushed into it. With deep

Table XXIX

Resistance of Case to Crushing Under a Brinell Ball

Double Treated Samples—High Carbon Case

Steel and Grain-Size	Case Depth Inch	Brinell Hardness from Applied Load of— 5,000 Pounds	10,000 Pounds	15,000 Pounds	20,000 Pounds	25,000 Pounds	Core Brinell	Load at First Crack Pounds
1020	.030	402	311	253	257		179	6,100
Grain-Size	.055	538	457	352	324		179	9,700
2-7	.066	590	532	467	421		179	13,900
	.074	680	550	506	451		174	14,000
1020	.037	454	386	352	288		192	8,000
Grain-Size	.062	563	431	434	396		192	14,000
2-3	.072	680	568	520	469		187	16,600
	.078	715	628	547	502		187	20,600
2315	.038	494	501	493	493		340	
Grain-Size	.064	563	568	547	538		321	
7	.068	590	568	547	526		340	
	.078	618	628	611	578		331	
2315	.033	436	431	404	387		241	
Grain-Size	.055	538	501	467	439		235	
2-3	.065	590	550	520	481		241	
	.085	590	587	547	526		235	
3115	.038	454	431	424	412		262	14,700
Grain-Size	.052	538	516	493	469		269	20,000
2-7	.065	563	550	520	493		248	
	.075	563	516	520	493		241	
3115	.032	402	365	338	294		196	9,900
Grain-Size	.048	494	444	434	372		196	16,300
2-6	.060	538	470	424	378		207	15,800
	.075	516	516	379	451		196	
4615	.035	418	386	386	339		248	10,100
Grain-Size	.048	516	444	424	396		207	15,300
7-8	.064	563	501	467	421		223	19,800
	.078	618	587	579	538		248	
4615	.040	454	418	395	366		235	14,800
Grain-Size	.050	516	470	424	396		228	15,600
3	.063	538	516	479	451		235	19,400
	.080	618	587	563	526		228	
2512	.036	484	484	506	493		401	
Grain-Size	.052	538	532	533	538		375	
6-7	.070	563	568	563	566		363	
	.080	538	587	579	578		363	
2512	.027	474	470	467	469		363	
Grain-Size	.048	538	532	547	526		375	
5	.065	538	550	563	551		363	
	.078	563	587	595	566		351	
6115	.034	418	407	395	366		235	12,600
Grain-Size	.056	516	457	454	412		235	
8	.060	538	550	520	493		248	20,000
	.078	494	532	493	481		248	
Krupp Gr.-7	.050	538	550	547	538		388	
4820 Gr.-7	.055	590	587	563	538		388	
4340	.054	648	650	647	623		514—400° Draw	
Grain 6-7	.060	494	516	506	502		444—800° Draw	

cases the load would become so widely distributed before reaching the core that the endurance limit of the core would not be exceeded. Fatigue failure would then take place in the case itself, starting either from the surface or at some intermediate layer at which the structure or internal stresses had created a zone of low endurance properties. The same reasoning could apply also to a lesser extent to the effect of other types of repeated loads.

Table XXX

Resistance of Case to Crushing Under a Brinell Ball

Pot-Quenched Samples—Low Carbon Case

Steel and Grain-Size	Case Depth Inch	Brinell Hardness from Applied Load of— 5,000 Pounds	10,000 Pounds	15,000 Pounds	20,000 Pounds	Core Brinell	Load at First Crack Pounds
1020	.024	336	311	298	278	187	5,700
Grain-Size	.042	494	407	352	324	196	9,000
2-7	.075	648	628	611	566	277	
1020	.025	494	501	454	460	286	10,000
Grain-Size	.044	563	532	520	502	277	10,000
2-3	.070	648	628	579	566	286	
2315	.027	474	470	467	469	364	
Grain-Size	.045	538	516	506	493	388	
7	.075	618	628	611	578	388	
2315	.030	474	444	434	412	293	15,000
Grain-Size	.050	538	516	493	493	311	20,000
2-3	.078	590	587	579	551	332	
3115	.023	387	386	377	351	286	10,200
Grain-Size	.045	563	550	533	514	302	20,000
2-7	.070	618–648	606	595	578	311	15,000
3115	.024	387	365	338	339	241	5,000
Grain-Size	.042	494	457	444	430	277	10,000
2-6	.073	648	628	519	551	262	20,000
4615	.024	387	365	352	351	311	10,000
Grain-Size	.040	474	444	424	396	262	10,000
7-8	.078	618	628	611	593	311	
4615	.025	454	431	413	396	255	5,000
Grain-Size	.048	516	470	467	430	248	15,000
3	.070	618	568	547	514	277	
2512	.023	454	470	468	460	415	
Grain-Size	.046	538	532	533	526	415	
6-7	.072	538	587	579	566	415	
2512	.028	454	457	454	451	364	
Grain-Size	.040	494	501	493	493	388	
5	.068	538	550	563	551	388	
6115	.024	418	386	361	351	262	9,700
Grain-Size	.044	538	501	467	451	293	15,800
8	.072	648	628	579	566	293	

Table XXXIII shows the effect of drawing temperature on the resistance to crushing of pot-quenched S.A.E. 4615 and 2512. The Brinell hardness of the 4615 is not lessened by drawing up to 500 degrees Fahr., and but little up to 600 degrees Fahr. on the deep case. The S.A.E. 2512 shows a continuous drop from the starting temperature of 400 degrees Fahr., and the deep case drops off at about the same rate as the shallow case.

The comparison of Brinell hardness numbers on uniform material but obtained by different loads is given in Table XXXIV. The steels were oil-hardened from 1525 degrees Fahr. and drawn as shown in the table. There was some variation in the 4640 steel itself and better comparisons can be made on the other two. There is a slight tendency for the hardness number to increase with the load.

Table XXXI

Resistance of Case to Crushing Under a Brinell Ball

Single Treated Samples—Low Carbon Case

Steel and Grain Size	Case Depth Inch	Brinell Hardness from Applied Load of 5,000 Pounds	10,000 Pounds	15,000 Pounds	20,000 Pounds	Core Brinell	Load at First Crack Pounds
1020	.025	373	311	309	278	179	8,000
Grain-Size	.045	474	396	370	339	187	12,500
2-7	.075	680	568	520	430	183	19,000
1020	.025	402	289	298	312	196	15,000
GrainSize	.046	494	431	404	378	207	17,300
2-3	.080	680	550	454	451	207	20,000
2315	.025	454	470	467	439	351	
Grain-Size	.048	563	550	547	538	375	
7	.075	563	568	563	551	340	
2315	.026	418	386	395	372	255	
Grain-Size	.055	516	501	493	469	286	
2-3	.075	590	568	547	538	269	
3115	.028	454	407	395	387	269	11,300
Grain-Size	.050	538	516	493	469	269	19,200
2-7	.072	590	568	563	538	269	
3115	.025	360	356	338	324	223	10,500
Grain-Size	.044	454	431	404	387	235	17,000
2-6	.064	590	550	520	481	235	19,600
4615	.026	325	319	304	306	255	15,000
Grain-Size	.050	494	431	424	396	269	16,000
7-8	.065	563	550	520	469	269	
4615	.024	348	338	322	312	235	10,000
Grain-Size	.040	436	407	404	396	228	17,300
3	.068	590	550	520	493	241	
2512	.030	454	484	479	481	388	
Grain-Size	.050	494	516	520	526	401	
6-7	.070	538	550	579	551	388	
2512	.025	454	470	467	469	388	
Grain-Size	.045	516	516	520	502	388	
5	.070	563	587	579	566	364	
6115	.031	454	457	444	405	277	11,300
Grain-Size	.050	563	516	493	469	277	
8	.078	590	587	579	538	286	

Oil Hardening Steels

To compare case hardened steels with medium carbon oil-hardened steels, a few tests were made on 4340, 4640 and 5140 of the analysis and grain-size given in Table I. Test specimens were of the same dimensions as previously described and they were hardened in oil after being heated in an electric furnace at 1525 degrees Fahr. Drawing temperatures were as shown in Tables XXXIV and XXXV. Results from the bend tests are given in Table XXXV and since never more than one crack occurred, the elastic limit has been listed instead. The elastic limits are of about the magnitude of the stress at the first crack in the double treated case hardened steels of the higher alloy content.

The strength calculated from the beam formula is considerably higher than that which would be obtained from an ordinary tensile

Table XXXII

Resistance of Case to Crushing Under a Brinell Ball

Double Treated Samples—Low Carbon Case

Steel and Grain-Size	Case Depth Inch	Brinell Hardness from Applied Load of 5,000 Pounds	10,000 Pounds	15,000 Pounds	20,000 Pounds	Core Brinell	Load at First Crack Pounds
1020	.022	348	302	279	248	183	5,600
Grain-Size	.038	418	365	345	318	179	10,900
2-7	.074	648	568	493	460	174	17,200
1020	.025	336	311	304	288	187	
Grain-Size	.045	538	407	386	351	187	11,300
2-3	.080	648	587	520	469	212	19,600
2315	.030	454	470	467	451	321	
Grain-Size	.060	538	550	533	526	352	
7	.075	590	587	579	566	340	
2315	.030	436	407	404	378	241	
Grain-Size	.047	494	457	434	412	241	
2-3	.068	563	550	520	502	248	
3115	.028	402	396	395	366	269	16,000
Grain-Size	.048	494	470	454	430	255	20,000
2-7	.068	590	568	547	514	262	
3115	.025	293	283	279	257	202	15,000
Grain-Size	.044	418	386	352	339	192	16,300
2-6	.065	538	484	444	396	187	
4615	.030	373	338	345	324	241	20,000
Grain-Size	.055	516	470	434	412	235	
7-8	.073	590	550	520	469	241	
4615	.027	387	365	345	324	235	14,400
Grain-Size	.050	454	407	404	372	223	18,800
3	.075	590	550	533	493	241	
2512	.030	464	457	467	439	363	
Grain-Size	.050	538	532	520	538	401	
6-7	.072	563	568	563	566	401	
2512	.028	474	470	479	469	388	
Grain-Size	.045	494	516	506	493	364	
5	.068	576	577	579	578	388	
6115	.030	360	363	352	339	228	14,000
Grain-Size	.048	494	457	434	412	262	
8	.075	474	470	434	405	223	

Table XXXIII

Effect of Drawing Temperature on Resistance of the Case to Penetration by a Brinell Ball

Pot-Quenched Samples

Steel and Grain-Size	Hardened Case Depth Inch	Drawing Temp. Degrees Fahr.	Brinell Hardness from Applied Load of 5000 Lbs.	10000 Lbs.	15000 Lbs.	20000 Lbs.	Core Brinell
4615	0.040	400	474	457	444	430	293
Grain-Size		500	474	457	434	412	321
7-8		600	373	365	352	339	262
		800	336	328	322	312	262
4615	0.075	400	563	532	520	514	302
Grain-Size		500	563	532	520	502	311
7-8		600	516	470	444	430	262
		800	418	386	370	366	255
2512	0.038	400	474	484	493	481	401
Grain-Size		500	454	457	454	460	401
6-7		600	402	407	413	412	401
		800	348	356	361	351	331
2512	0.075	400	516	532	533	526	415
Grain-Size		500	494	501	493	493	415
6-7		600	454	457	454	460	388
		800	380	386	395	378	321

Table XXXIV
Influence of Load on Brinell Hardness Number

Steel and Grain-Size	Drawing Temp. °F.	Brinell Hardness from Applied Load of— 3000 Kgm.	5000 Lbs.	10000 Lbs.	15000 Lbs.	20000 Lbs.	25000 Lbs.	Brinell at Center 3000 Kgm.
4340 Grain 6-7	400	534	538	568	579	578	586	514
	800	429	402	431	444	451	431	430
4640 Grain 8	400	514	516	550	547	538		477
	500	477	454	501	493	514		461
	600	460	454	457	479	493		444
	800	375	373	386	386	378		401
5140 Grain 3	400	555	538	587	611	623		555
	500	514–534	516	550	579	578		534
	600	514	494	516	533	538		514
	800	430–444	418	444	454	451		444

Table XXXV
Results from Bend Tests on Oil Hardening Steels (Not Carburized)

Steel	Elastic Limit Pounds per Sq. In.	Def. at E. L. Inch	Maximum Pounds per Sq. In.	Max. Def. Inch	Drawing Temp. °F.	Rockwell of Surface	Brinell at Center	Fracture
4340	318,600	.162	533,400	.725	400	52-54	514	Fine cryst. and part silky.
4340	267,550	.133	369,800	.750	800	43-44	430	No break.
4640	270,400	.141	474,000	Over ¾	400	45-51	477	Mostly fine cryst.
4640	273,250	.137	446,500	Over ¾	500	47-49	461	Fine crystalline.
4640	256,200	.128	381,400	Over ¾	600	45-48.5	444	Fine crystalline.
4640	219,100	.108	319,000	Over ¾	800	36-40	401	No break.
5140	304,500	.153	342,000	.182	400	54-55	555	Fairly fine cryst.
5140	298,700	.143	298,700	.143	500	53	534	Slightly coarser.
5140	293,000	.143	293,000	.143	600	50-51.5	514	Same
5140	259,000	.125	376,000	Over ¾	800	43-45	444	Fine crystalline.

Table XXXVI
Izod Properties of Oil Hardening Steels

Steel	Foot-Pounds Notched	Foot-Pounds Unnotched	Drawing Temp. °F.	Rockwell Hardness	Fracture
4340	17	120	400	53	Very fine crystalline
4340	18	120	800	44	Very fine crystalline, part silky
4640	10	120	400	53	Very fine crystalline
4640	7	120	500	51	Very fine crystalline
4640	9	120	600	48	Very fine crystalline
4640	25	120	800	40	Half silky
5140	2½	39	400	55	Fairly fine crystalline
5140	2	14½	500	52	Slightly coarser
5140	1½	10	600	50	Same
5140	12	120	800	44	Fairly fine crystalline

test. The S.A.E. 4640, probably because of its very fine-grained abnormal structure, had rather poor hardening capacity and soft spots were found below the average hardness values given in the table. It might also be mentioned that it was more difficult to clean

the scale from those samples. Apparently the greater the hardness of a quenched specimen, the more readily the scale is removed. This may be a partial explanation of the greater difficulty in cleaning the case hardened nickel steels, as their surface hardness also was lower than some of the other steels. Scale is removed more readily before tempering than afterwards.

The impact results given in Table XXXVI show in most cases a greater ratio between notched and unnotched values than do most of the case hardened specimens. This means that scratches, deep marks, sharp corners, etc., on oil-hardened parts are even more to be avoided than on case hardened parts, since the former may lose a greater percentage of their possible maximum shock resistance by such imperfections. The unnotched specimens, except from S.A.E. 5140, did not break under the full capacity of 120 foot-pounds.

Minimum impact values do not correspond with minimum drawing temperatures. The fracture appeared somewhat coarser at the intermediate drawing temperatures, at which also the lowest impacts occurred.

Wear Tests

It was planned to make wear tests on samples from each steel, grain-size, case depth and heat treatment, but after running about 200 samples the tests were abandoned as being too erratic. When a lubricating oil was used, no appreciable wear occurred in many of the samples even under high loads and several hours running time. When wear did occur, it was very rapid due to the block becoming tilted and concentrating the load at one point or due to seizure of metal surfaces due to the breaking down of the oil film. In order to get greater wear, a lubricant consisting of 1 per cent soluble oil in water was tried. All lubricated samples were run for four hours with a 50-pound load on the lever arm.

A considerable number of samples were run dry in which case the time was only ten minutes and the load two pounds. The dry samples developed considerable heat, the rapid wear probably being partly due to tempering and softening of the case. On a time-weight basis, water with as little as 1 per cent soluble oil increased wear resistance several hundred fold over no lubrication, in fact higher percentages of soluble oil proved to be an excellent lubricant. The tests run with lubricants were apparently more of a test of the film strength of the lubricant than of the wear resistance of the steel.

Because of erratic results and failure to get reasonable checks

Table XXXVII
Tensile Properties of the Core

Samples Quenched from 1700° F. and drawn at 325° to Correspond to Pot Quench.

Steel and Grain-Size	Proportional Limit Lbs. Per Sq. In.	Johnson Limit Lbs. Per Sq. In.	Max. Lbs. Per Sq. In.	Per Cent Elongation	Per Cent Reduction of Area	Brinell
1020 Grain 2-7	36,500	45,000	102,000	18.28	54.4	235–241
1020 Grain 2-3	32,500	40,000	109,000	14.06	39.4	248–255
2315 Grain 7	58,500	75,000	166,500	14.45	44.7	341–352
2315 Grain 2-3	48,500	75,000	155,000	13.28	45.2	321–331
3115 Grain 2-7	41,000	60,000	127,000	18.46	52.2	269
3115 Grain 2-6	30,500	45,000	108,000	23.44	63.6	229
4615 Grain 7-8	40,000	55,000	115,250	25.64	64.3	241–255
4615 Grain 3	42,500	55,000	133,900	15.24	54.9	302
2512 Grain 6-7	70,000	110,000	190,900	13.67	50.0	401
2512 Grain 5	78,000	120,000	197,000	14.45	50.7	401–415
6115 Grain 8	47,500	70,000	131,250	20.70	61.5	277–285
Krupp Gr. 7	62,500	75,000	157,500	17.58	64.9	331–352
4820 Gr. 7	49,750	65,000	142,200	18.75	54.9	302–311
Cr-Ni-Mo. 7 Grain 4	60,000	80,000	168,000	15.63	56.0	363
Cr-Ni-Mo. 8 Grain 4 *	80,000	135,000	225,000	12.5	47.5	477

*Heat Treated after machining.

on the same sample, no tabulated results will be given. The most that can be said is that wear resistance in general tends to coincide with the degree of file hardness. From the average of all the methods, some of the water-hardened plain carbon steels showed the least wear and the nickel steels usually showed the most whether dry or lubricated.

Core Properties of the Case Hardening Steels

As mentioned earlier, all specimens were put through a heat treating cycle to approach as nearly as possible the condition of the core in a case hardened part. Obviously the conditions of internal stress between case and core cannot be duplicated and the core properties given may not mean a great deal in predicting properties of the finished part, as may be seen by comparing them with the results from the bend tests. The specimens were treated as 7/8-inch

Table XXXVIII

Tensile Properties of the Core

Steel and Grain-Size	Proportional Limit Lbs. Per Sq. In.	Johnson Limit Lbs. Per Sq. In.	Max. Lbs. Per Sq. In.	Per Cent Elongation	Per Cent Reduction of Area	Brinell
Single Treated Samples						
1020 Grain 2-7	30,500	30,000	83,000	32.42	62.2	166–170
1020 Grain 2-3	31,000	30,000	91,750	24.61	45.7	192
2315 Grain 7	55,000	65,000	148,600	18.36	50.7	321
2315 Grain 2-3	39,000	50,000	130,750	19.14	58.9	269
3115 Grain 2-7	32,500	40,000	124,600	23.44	50.6	255
3115 Grain 2-6	30,000	30,000	100,250	28.13	59.4	202–212
4615 Grain 7-8	40,000	50,000	113,500	27.34	62.2	235–241
4615 Grain 3	40,000	55,000	124,600	20.31	61.2	269
2512 Grain 6-7	70,000	105,000	186,900	15.63	54.3	401
2512 Grain 5	70,000	115,000	197,250	15.23	56.7	415
6115 Grain 8	45,000	65,000	126,500	23.44	59.3	255–269
Krupp Gr. 7	57,250	70,000	151,000	17.97	64.5	311–341
4820 Gr. 7	47,000	60,000	135,000	21.09	59.5	285–302
Cr-Ni-Mo. 7 Grain 4	45,000	65,000	149,500	15.63	62.1	321
Cr-Ni-Mo. 8 Grain 4 *	85,000	135,000	231,000	14.06	49.5	477

*Heat Treated after machining.

rounds in most cases and not to exceed 1 inch in thickness in the others. Machining was done after heat treatment except in one or two cases where, because of the high hardness after treatment, it was necessary to heat treat the specimens as such.

Considerable difficulty was encountered in determining the proportional elastic limit. Two methods were tried, one by using an electrical extensometer reading to 0.0002 inch and applying the load in increments, and the other by a self-recording instrument attached directly to the specimen itself. Neither proved highly satisfactory. It was found that in plotting stress and strain readings, a break in the curve might occur after which another series of readings would lie in a straight line, that is, the strain was proportional to the stress, but at a different slope than the lower part of the curve. The proportional limits given are lower than those frequently reported but apparently the more sensitive the measurement, the lower the elastic limit

Table XXXIX
Tensile Properties of the Core

Steel and Grain-Size	Proportional Limit Lbs. Per Sq. In.	Johnson Limit Lbs. Per Sq. In.	Max. Lbs. Per Sq. In.	Per Cent Elongation	Per Cent Reduction of Area	Brinell
	Double Treated Samples					
1020 Grain 2-7	33,750	35,000	82,200	33.98	69.1	170–179
1020 Grain 2-3	34,000	35,000	86,150	31.25	59.1	179–183
2315 Grain 7	35,000	35,000	125,250	29.29	54.1	255–269
2315 Grain 2-3	26,000	35,000	113,000	28.13	57.0	229–235
3115 Grain 2-7	30,250	40,000	116,500	28.13	52.0	241
3115 Grain 2-6	27,000	30,000	99,750	32.42	60.3	197
4615 Grain 7-8	31,000	40,000	107,500	32.03	64.2	223
4615 Grain 3	31,000	40,000	108,500	29.30	56.6	229
2512 Grain 6-7	35,000	60,000	153,000	15.63	54.8	331–341
2512 Grain 5	41,500	55,000	147,000	17.57	55.8	302–341
6115 Grain 8	35,000	40,000	112,000	27.34	56.6	217–229
Krupp Gr. 7	65,000	75,000	156,000	18.75	65.7	341
4820 Gr. 7	45,000	60,000	134,250	21.09	60.6	302–311

becomes. The maximum strength of any steel irrespective of analysis, whether alloy or plain carbon, is closely proportional to the Brinell hardness provided no defects exist in the steel. The proportional limit showed no such relationship and the writer doubts if the elastic limit as usually measured gives any useful information about the steel. The true elastic limit, if it could be accurately measured, would probably coincide with the endurance limit. The results of the tensile tests are given in Table XXXVII to XXXIX. The Johnson limit has been included also as it appeared to be more definite than the proportional limit. The Brinell hardness was taken at the center of the cross section cut from the threaded end of the plug after it was pulled.

Core impacts are given in Tables XL to XLII. All the unnotched specimens exceeded 120 foot-pounds. The Brinell hardness range given is that obtained from three impressions on each specimen. Of the steels showing distinct differences in grain-size, the fine grained S.A.E. 1020 and 4615 have higher impact values than the corresponding coarse-grained heats, but those on the coarse-grained 2315 are higher just as they were in the case hardened condition. The

Table XL
Core Impact Values

Samples Treated to Correspond to a 1700° F. Pot Quench and 325° Draw

Steel and Grain-Size	Foot-Pounds	Ft.-Lbs. Average	Brinell Hardness	Fracture
1020 Grain 2-7	40, 37, 49, 43	42	214–242	⅛ medium cryst. Rest silky.
1020 Grain 2-3	16, 14, 19, 17	16	242–276	Mixed silky and fairly fine cryst.
2315 Grain 7	33, 33, 34, 35	34	319–343	Silky.
2315 Grain 2-3	30, 32	31	330	Silky with scattered grain.
3115 Grain 2-7	44, 41, 43, 41	42	242–276	⅛ fine cryst. Rest silky.
3115 Grain 2-6	84, 89, 77, 84	83	201–234	Silky.
4615 Grain 7-8	93, 98, 97, 97	96	220–242	Silky.
4615 Grain 3	29, 32	30	276–296	¾ fine cryst. Rest silky.
2512 Grain 6-7	33, 35, 30, 30	32	370	Silky.
2512 Grain 5	27, 30, 28, 31	29	370–385	Silky.
6115 Grain 8	85, 83, 77, 76	80	242–276	Silky.
Krupp Gr. 7	78, 77, 73, 79	77	296–307	Silky.
4820 Gr. 7	36, 35, 46, 44	40	296–343	Silky with scattered grains.
Cr-Ni-Mo. Grain 4	42, 43	42	343–370	Silky.
Cr-Ni-Mo. Grain 4 *	34, 34, 32, 32	33	400–418	Silky.

*Heat treated after machining.

coarse-grained 2315 is softer than the fine-grained, which would account for at least part of the difference. All three types, however, show a greater improvement in the coarse-grained steels when double treated than do the fine-grained heats. The tendency for most of the steels is to show higher impacts when double treated, the 6115 being the most noticeable exception. There is further evidence as shown by the coarse-grained single-treated 1020 and 3115 that steel hardened from within the critical range has low impact values. The high impact values of the fine-grained 4615 compares with its standing among the steels as case hardened. There appears to be a little more relationship between core impacts and impacts on the case hardened parts than there is between core tensile properties and beam strength of case hardened specimens. The relationship does not exist in all cases, as may be shown by 6115.

Summary and Conclusions

The results of various tests determining the physical properties

Table XLI
Core Impact Values

Single Treated Samples

Steel and Grain-Size	Foot-Pounds	Ft.-Lbs. Average	Brinell Hardness	Fracture
1020 Grain 2-7	58, 52, 64, 66	60	152–174	⅓ fine cryst. Rest silky.
1020 Grain 2-3	12, 14, 14, 14	13	170–179	Mostly medium cryst.
2315 Grain 7	46, 52, 47, 46	48	286–296	Silky.
2315 Grain 2-3	56, 52, 58, 58	56	242–276	Mostly silky.
3115 Grain 2-7	40, 54, 47, 55	49	242–258	Mostly silky.
3115 Grain 2-6	64, 65, 70, 66	66	190–201	Silky with scattered grains.
4615 Grain 7-8	86, 88, 79, 97	87	214–227	Silky.
4615 Grain 3	44, 41, 31, 44	40	250–267	¾ fine cryst. Rest silky.
2512 Grain 6-7	32, 34, 29, 33	32	356–370	Silky.
2512 Grain 5	30, 28, 25, 27	27	370–385	Silky.
6115 Grain 8	78, 87, 75, 86	81	242–258	Silky.
Krupp Gr. 7	76, 75, 53, 54	65	276–318	Silky.
4820 Gr. 7	68, 69, 53, 56	61	267–296	Silky.
Cr-Ni-Mo. 7 Grain 4	52, 64	58	296–330	Mostly silky.
Cr-Ni-Mo. 8 Grain 4 *	33, 33, 35, 32	33	436	Silky.

*Heat treated after machining.

Table XLII
Core Impact Values

Double Treated Samples

Steel and Grain-Size	Foot-Pounds	Ft.-Lbs. Average	Brinell Hardness	Fracture
1020 Grain 2-7	101, 102, 103, 88	98	148–165	Silky.
1020 Grain 2-3	47, 47, 53, 58	51	161–179	½ fine cryst. Rest silky.
2315 Grain 7	52, 56, 58, 61	57	242–267	Silky.
2315 Grain 2-3	76, 76, 77, 77	76	201–220	Silky.
3115 Grain 2-7	60, 62, 55, 58	59	207–214	Silky.
3115 Grain 2-6	79, 77, 81, 79	79	179–190	Silky.
4615 Grain 7-8	92, 95, 92, 93	93	195–207	Silky.
4615 Grain 3	74, 75, 69, 68	71	195–214	Silky with scattered grains.
2512 Grain 6-7	40, 45	42	250–307	Silky.
2512 Grain 5	45, 54, 61, 56	54	242–296	Silky.
6115 Grain 8	57, 60, 52, 56	56	201–214	Silky.
Krupp Gr. 7	58, 56, 73, 81	67	296–318	Silky.
4820 Gr. 7	62, 61, 56, 61	60	276–318	Silky.

of both the case and core of a number of case-hardening steels as well as a few oil-hardening steels are given in this paper. The object was to compare the properties of the various steels, one with another, and also to determine the influence of grain-size, depth of case, carbon content of the case, and method of heat treatment on any one type of steel. The writer had in mind chiefly the suitability of the steels for gears but the results are applicable to other uses as well. A fine- and coarse-grained heat of each of S.A.E. 1020, 2315, 3115, 4615, and 2512, and a fine-grained heat of 6115 were used in most of the tests, but some results on Krupp, 4820, 4340, 4640, 5140 and two lower carbon chromium-nickel-molybdenum steels are reported. The properties determined were the microstructures in both the carburized and heat treated condition, the rate of carburizing, the hardness of the case and core, the strength of a case hardened part loaded as a beam, the impact strength, the amount of distortion in carburizing and hardening, the resistance of the case to crushing under heavy loads, and its wear resistance. Hardness, tensile and impact tests were made on the core.

The microstructures of the cases are best described by an examination of the photographs themselves. The appearance of the hardened structures will be found listed in tables. The pot-quenched samples, including S.A.E. 1020, which was the only steel quenched in water, all contained austenite in the case, except some of the shallow cases and those of lower carbon content. The higher alloys retained some austenite in the single treatment but none was visible in the double-treated samples. The structures were such as to be expected from the type of steel, grain-size, carbon content and treatment given. Carburized and slowly-cooled samples from every group of test specimens were examined for character and depth of case. No specific investigation was made to determine the rates of carburizing among the various steels but where conditions of carburizing were identical, comparisons were made. Only a limited number of such comparisons were made, but from these there was shown some tendency toward deeper cases in the coarse-grained steels but this was not without exception. As to the rate among types of steel, S.A.E. 6115 carburized the fastest followed closely by 3115, then 1020, 2315 with 4615 and 2512 the slowest and at about the same rate. The other steels were not tested at the same time and direct comparisons cannot safely be made.

The general tendency of the strength of the case as determined

by a bend test was to follow the strength of the uncarburized steel, but there were many exceptions to this. Theoretically, because of the greater hardness and strength of the case, the deeper the case the greater should be the strength of the case-hardened bar. Using the load at the first crack of the case as a measure of its strength, with the single exception of the fine-grained S.A.E. 4615, the strength dropped off with increase of case depth when pot-quenched, especially in such steels as 2512 with higher alloy content. This is probably due to internal stresses set up in hardening as shown by the fact that higher drawing temperatures on some pot-quenched samples increased the strength of all samples, but more on the deep cases to make them stronger than the shallow cases. The strength still decreased with case depth on the higher alloys when single-treated, but increased on the lower alloy steels. Strength increased with case depth on the double-treated samples. The strength of some of the steels with higher alloy content and deep cases was nearly doubled over that as pot-quenched when double-treated, showing that the former is not a suitable treatment for such steels if high static strength is desirable. Coarse-grained steels were more improved by a double treatment than were the fine-grained. Low carbon cases appear to be less influenced by case depth and heat treatment than do high carbon cases. The Brinell hardness of the core is not necessarily a measure of the strength of a case-hardened part. The number of cracks occurring in the case before final failure in the bend test had no connection with the strength or with the toughness as determined from impact results except when influenced by case depth. The thin cases did develop more cracks and gave higher impact values than did the deep cases. Most of the pot-quenched samples developed the most cracks in the case and the coarse-grained more than the fine.

Impact tests were made on case-hardened bars in both the notched and unnotched condition. There was no uniform relationship between notched and unnotched values. The nickel steels were high in unnotched values but were affected more by the notch than the other steels, indicating that scratches, sharp corners, etc., more adversely affect these steels in service and result in the loss of most of their superior shock-resisting properties. In general the single-treated samples gave the lowest impact values and the double-treated samples the highest. This is considered to be due in part at least, to the lack of uniformity of the core in the single-treated samples. The grain-size and uniformity of structure of the core, rather than its hardness,

has been found to have the most influence on impact values. Greater improvement by double treatment occurs in coarse-grained steels than in fine. The fine-grained heats were in general higher in impact than the coarse-grained heats but the reverse was true with S.A.E. 2315. The S.A.E. 6115, although the finest-grained steel of those tested, developed some of the lowest impact results. The carbon content of the case had but little effect on the impact values. Case depth has the greatest influence on impact, the impact values decreasing with increase of case depth. The fractures of both the case and core were similar to the corresponding bend tests, showing that rate of breaking has little influence on the appearance of the fracture. The appearance of the fracture was not affected by the notch. The notch was made before case hardening.

The Rockwell hardness of case-hardened parts is dependent upon the per cent of martensite in the case. High carbon contents, high quenching temperatures such as pot-quenching, and increasing alloy content increase the amount of austenite and lower the hardness readings. Decarburization during reheating also lowers the Rockwell hardness. The file hardness does not follow the Rockwell hardness as in general the pot-quench samples were hardest to the file and the double-treated samples the softest, not only on the surface but underneath the decarburized layer. Although low-hardening temperatures give the highest Rockwell readings, higher hardening temperatures tend to give greater resistance to the file. This has not been found to be the case with plain carbon steel when water-hardened, probably due to the greater tendency to form steam pockets, rather than the temperature effect on the steel itself. The water-hardened plain carbon steel was hardest to the file, with 3115, 4615, and 6115 next. The higher nickel steels were least hard to the file. Scale adherence also was greatest to the nickel steels and may be connected with hardness obtained in the quench.

Distortion tests showed the greatest change in shape to occur in the water-hardened S.A.E. 1020. Not much difference was observed among the alloy steels with the test blocks used. Great regularity was observed in the changes which did occur when conditions were kept identical. Case depth was the greatest factor causing differences in distortion. Case depth, however, had little influence when the parts were pot-quenched, probably due to the greater influence of the core.

The practical value of case-crushing tests is doubtful. Light

loads far too small to cause any general sinking of the case into the core produced a Brinelling action on the surface which would destroy the parts for such use as gears. Spalling, pitting, and other case failures are believed due to fatigue starting in the core below the case when the case is thin, or at the surface or some intermediate zone of internal stress when the case is deep. Brinell hardness of the case was influenced by case depth, hardness of the core, and the amount of austenite in the surface layers.

The elastic limits of the oil-hardening steels in bending were comparable in value with the first crack in the case of the double-treated case-hardened steels of higher alloy content. The impact values of notched specimens were comparable with the notched case-hardened specimens but, with the exception of 5140 which was low both notched and unnotched, the unnotched values were higher. Oil-hardened steels are, therefore, even more adversely affected in their shock resistance by surface defects than are case-hardened steels.

Wear-test results were not consistent. When a lubricant was used, the result was mostly a test of its film strength and when run dry, samples heated and wore rapidly. Indications were that wear-resistance follows file-hardness. Nickel steels appear to have the least resistance to wear.

Tensile and impact tests were made on pot-quenched, single and double treated samples. Tensile strength follows closely with Brinell hardness and is highest on the higher alloy steels such as 2512. There is more relationship between impact values of the core and the case-hardened samples than in strength of bars, but there are exceptions, especially 6115. Fine-grained steels had higher impact values than the coarse except 2315. Double-treated samples except 6115, had the highest impact values.

Highest load carrying capacities are obtained with the higher alloy steels such as 2512, Krupp and 4820, if properly treated. Especially if deeply cased, they should not be pot-quenched. For high shock resistance, nickel steeels were best when unnotched, but lost a great deal of their advantage when notched, as did also the oil-hardening uncased steels. S.A.E. 4615 steel had the highest average for notched impacts. The properties of that steel when fine-grained, appear to make it ideal for pot-quenching. The best properties of S. A. E. 6115 were developed by a single quench. S.A.E. 1020 has low strength and shock resistance. S.A.E. 3115 has average properties in strength and apparently good wear resistance. Fine-grained,

case-hardened steels were superior in both strength and impact with the exception of the impact of coarse-grained 2315.

Other factors than those listed may be of importance for certain types of service. For example, dynamometer tests have shown that pot-quenched pinions, even though the strength may be lower, are less affected by high operating temperatures resulting from partial lubrication. Pot-quenched cases are softened less than reheated cases by equal drawing temperatures and scuff less in service. Those results will be made the subject matter of another paper.

Acknowledgments

The author wishes to thank the Timken Steel and Tube Company for supplying most of the steel used in these tests and state that the work was made possible through the facilities of the Timken-Detroit Axle Company.

DISCUSSION

O. E. Harder:[1] In discussing steel 2512, we note a marked increase in the strength with high drawing temperatures. I should like to inquire to what extent that increase in strength is due to a transformation of retained austenite to martensite or troostite on tempering.

In connection with the observation that these carburized steels fail in fatigue at the base of the carburized zone, it is of some importance to mention that a similar phenomenon has occurred in the case of nitrided steel, the fatigue test showing that the failure starts at the base of the nitrided case.

O. W. McMullan: Examination of Fig. 20 shows that the maximum strength was found at about the maximum hardness, which would indicate that that was the point at which most of the austenite was transformed to martensite, or troostite, as the case may be, so that change in strength did go along with change in microstructure to give the maximum strength where most of the austenite was first transformed.

Hyman Bornstein:[2] I think the author should be complimented on this excellent paper. It is just the type of so-called "practical paper" that the Society should have at its meetings. It is full of useful and practical information. I would like to ask the author about Fig. 22, which shows distortion. It is a little complicated, and I was wondering whether it showed that there was more variation in distortion on the straight carbon steel with change in case depth than with the alloy steels.

O. W. McMullan: Yes. The straight carbon steel varied more.

Hyman Bornstein: And there is also a wider range of distortion on the plain carbon steels?

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[2]Metallurgist, Deere and Company, Moline, Ill.

O. W. McMULLAN: Yes, particularly on those quenched direct from the carburizing temperature.

J. FLETCHER HARPER:[3] I wish to compliment Mr. McMullan on his paper. We have done similar work, paralleling his results very closely. I wish to confirm one suspicion that he has put forward, namely that the fracture takes place between the case and the core. With our bending tests, we applied a microphone to the test bar and with high amplification we got a crushing, grinding sound which, to us, intimated a tearing apart at that section of the test piece prior to any definite crack upon the surface. We considered this sound to be a fracture taking place between the case and the core.

J. P. WALSTED:[4] I would like to ask Mr. Harper if he confirmed that conclusion by stopping the test at the time he got the breaking sound and cutting a cross section to find if fracture had occurred. It seems to me that that would be a rigorous proof of whether or not crushing did occur.

GORDON WILLIAMS:[5] I would like to ask, in connection with the distortion charts, whether the measurements of the pieces that were not directly quenched after carburizing showed the majority of distortion to be from the carburizing treatment—that is in the soft condition, after carburizing.

O. W. McMULLAN: On the single and double treated samples as in Figs. 22-24—the horizontal lines denote the change after carburizing only, and the slanting lines the changes occurring during the hardening, so the end of the horizontal line represents the position of the corners after carburizing and cooling in the box.

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[4]Instructor, department of metallurgy, Massachusetts Institute of Technology, Cambridge, Mass.

[5]Metallurgist, Cleveland Tractor Co., Cleveland.

THE INFLUENCE OF NONMETALLIC INCLUSIONS UPON THE PRECIPITATION OF PRIMARY CEMENTITE IN HYPEREUTECTOID STEELS

By E. G. Mahin and E. F. Lee

Abstract

The effect of nonmetallic inclusions upon the segregation of primary ferrite in hypoeutectoid steels has been investigated and reported in earlier papers by many authors but the influence of such inclusions upon the segregation of primary cementite in hypereutectoid steels has received comparatively little attention, in the past. In this investigation, high carbon steel was produced by cementation of ingot iron and wrought iron, nearly carbon-free materials which, in the process of manufacture, must have given little opportunity for segregation of traces of dissolved impurities, the influence of such segregation upon phase changes being thus eliminated.

Pronounced cementite segregation resulted. An explanation for this action is proposed, taking into consideration the analogous effect in connection with ferrite segregation, and certain suggestions are offered in connection with the effects of carburization of "abnormal" steels.

ON account of their importance in connection with strength characteristics, corrosion problems and machining qualities of metals and alloys, the problems presented by the presence of foreign bodies of nonmetallic character have received a large share of attention from investigators in the field of metallurgy. Probably these problems are of greater importance in ferrous metallurgy because ferrous alloys find their widest use where mechanical strength is a major item in their selection.

An extensive discussion of this subject, including a comprehensive bibliography, was published in 1931 by Benedicks and Löfquist in their book, "Non-Metallic Inclusions in Iron and Steel." The text of this book covers practically every phase of the subject upon which research papers have been published and no attempt

A paper presented before the Sixteenth Annual Convention of the Society held in New York City, October 1 to 5, 1934. Of the authors, E. G. Mahin is head of the Department of Metallurgy, University of Notre Dame, and E. F. Lee is a student of the same department, class of 1934. Manuscript received July 2, 1934.

will be made, in the present paper, to repeat the survey of the literature, the citations being limited to such papers as bear upon the phases of the problems here discussed.

Some years ago the senior author became interested in the much discussed question as to the influence of nonmetallic inclusions upon the segregation of the ferrite phase in hypoeutectoid steels. The association of inclusions and ferrite had been noticed, probably ever since the introduction of microscopic methods for metal studies. Whether this relation was in the nature of a coincidence, the location together of primary ferrite and inclusions resulting from a common cause, or whether the presence of the inclusions in certain localities was a fundamental cause of the separation of the ferrite phase in these regions, was a question upon which opinion divided but the inclination was toward the latter view, with a satisfactory explanation still lacking.

Beginning in 1919, six papers dealing with this phase of the question were published by the senior author and his students (1), (2), (3), (4), (5), (6).* The general argument contained in these papers not only supported the view that nonmetallic inclusions cause premature separation of primary ferrite around them, but presented also a tentative explanation of this action. It was assumed that a very shallow zone of metal, in the immediate vicinity of the inclusion, is contaminated by elements or compounds derived from the inclusion and in solid solution. When a steel is heated into the γ-temperature zone and then slowly cooled, the dissolved material causes the first separation of primary ferrite to occur in this region and, thereafter, ferrite will normally crystallize upon nuclei so formed.

Experimental evidence was presented to support this view and it was shown that almost any chemical element, existing in low concentration in solid solution but segregated in certain localities, will cause such ferrite segregation. Stead, in several papers (7), has supported his suggestion that phosphorus segregation is alone responsible for ferrite segregation and his analysis of the behavior of a steel cooling from above the liquidus and leaving a high phosphorus content, as well as most of the inclusions, in the portions of the liquid phase last to freeze, is sufficiently convincing. It has been noted, however, that after destroying phosphorus banding in a rolled steel by heating for 72 hours, co-segregation of inclusions and primary ferrite was as evident as ever (1). It has further been shown (2)

*The figures appearing in parentheses refer to the bibliography appended to this paper.

that artificially produced segregation of sulphur, manganese, silicon, aluminum, copper, zinc, chromium, tungsten, nickel and titanium also caused ferrite segregation, in some cases in a very pronounced manner.

The mechanism of the phase changes occurring as austenite cools slowly through the transformation range is similar to that of other binary solutions, liquid or solid, in systems forming eutectics. Supercooling always occurs, even though sometimes to a small extent. The iron-carbon solid solution cools past the temperature of γ-α equilibrium (or of γ-iron-cementite equilibrium) and the temperature at which the primary phase, ferrite or cementite, begins to crystallize is somewhat indefinite, depending as it does upon composition, rate of cooling and the possible presence of foreign bodies or of traces of dissolved additional elements. This spread between A_3 and Ar_3 (or between A_{cm} and Ar_{cm}) constitutes the thermal hysteresis gap of the upper portion of the transformation range.

In attempting to understand the effect of a third component, added to an existing binary solution, attention is too often entirely focused upon the equilibrium diagram for the ternary system so formed. Naturally the expected effect of the first additions of the third component is to lower the temperature at which a given transformation is to take place, but it must be remembered that the new temperature, as represented on the constitutional diagram, is for equilibrium between the phases existing above and below this temperature.

In the present case we cannot ignore another effect of the presence of the third component, especially as this effect may, in some cases, be of greater importance than the one just mentioned, as well as being entirely independent of it. The effect referred to is the narrowing of the hysteresis gap, due to starting the breakdown of the metastable solid solution at a higher temperature than would otherwise be the case.

As a convenient illustration, one may consider the effect of the presence of a very small amount of sulphur in an iron-carbon alloy. From the standpoint of the iron-carbon-sulphur diagram, the intersection of the temperature vertical with the surface separating the γ and the γ-α zones would be lowered but very slightly by the presence of a few hundredths of a per cent of sulphur, such as would be found in any even moderately good steel or iron. "Slow cooling" is an ambiguous expression, at best, but if we limit its use to include any

cooling rate that will permit the formation of normal lamellar pearlite, the separation between A_3 and Ar_3, for a hypoeutectoid steel, might be as much as 15 to 25 degrees Cent. If it may be assumed, merely for illustrative purposes, that the traces of sulphur in a steel would lower the equilibrium temperature, A_3, one or two degrees, but would raise the actual temperature of the beginning of phase change by five degrees, then it is easy to see that ferrite formation would begin at a higher temperature than would have been the case in a pure iron-carbon alloy having the same carbon content.

The actual figures used in the above illustration are not essential to the reasoning. The only assumption that has been made is that the trace of the third component raises the temperature represented by Ar_3 to a greater extent than it lowers A_3, and this assumption is in accord with the known behavior of metastable solutions.

Let us consider now the case of an inclusion of manganese sulphide. It has been claimed by Giolitti and Tavanti (8) that they have experimentally proved that slag and other similar inclusions are not soluble in solid iron. This conclusion is not only in direct contradiction to a great mass of experimental evidence, but it also is not in accord with theoretical reasoning, since there is, theoretically, no such condition as zero solubility. Their experimental method was based upon microscopic measurements of the sizes of inclusions in steel, annealed and quenched from the liquid state, respectively. Considering the fact that only average sizes could be determined in this way and, even then, only with a rather large factor of error, such a method could not be conclusive unless large differences in solubility existed—a statement which we believe nobody claims to be true. On the other hand, it has been proved beyond a doubt (4) that the zone of metal surrounding manganese sulphide inclusions does contain appreciably higher sulphur concentrations (and, no doubt, of manganese) than does the metal farther removed from the inclusion.

It may be proper to repeat here, in this connection, a statement that has been made in a previous paper, that this is another reason why the analytical determination of sulphur in steel is of such variable significance. Only total sulphur is usually determined and there is, from the chemical analysis alone, no indication whatever as to the relative amounts of this sulphur existing as glassy manganese sulphide and as sulphur dissolved in the metallic phases. The former is harmful in a purely mechanical way and probably is not very im-

portant, from the standpoint of strength characteristics, because it usually occupies a smooth and well rounded cavity. But the dissolved sulphur is, of course, extremely objectionable. Even then, if the dissolved element was uniformly distributed throughout the mass, it would seem that its effects would be negligible, or nearly so.

There is no present method for determining what concentration of sulphur does exist in the metal shell in immediate contact with an inclusion of manganese sulphide, but it may easily be possible that this concentration may sometimes be considerably higher than the figure given by analysis of the sample of steel as the chemist determines it. If even a thin shell of metal surrounding the inclusion is decidedly "short", on account of dissolved sulphur, we are then dealing with a potential starting point for failures under static, impact or repeated stresses.

If it is correct to assume that the presence of a small concentration of sulphur, for example, may diminish supercooling of austenite to a greater extent than it lowers the equilibrium temperature of the phase change, then the shell of high sulphur austenite in immediate contact with the sulphide inclusion will experience the beginning of α-iron formation, in the cooling steel, before any primary ferrite forms elsewhere. Since, at all temperatures down to Ar_1, the austenite is in a metastable condition, ferrite will normally crystallize, to a considerable extent, upon the nuclei so formed and the familiar microstructure will be noticed in the prepared section. If there are bands or threads of inclusions, the longitudinal section will exhibit ferrite banding, whether or not phosphorus banding is present. And the same reasoning will, of course, apply to other segregated elements in any cooling hypoeutectoid steel.

Hypereutectoid Steels

If the validity of this hypothesis is admitted, there still remains to be tested its application to steels in which the separating primary phase is cementite, rather than ferrite.

Stead (7), in explaining his conception with regard to the effect of segregated phosphorus in promoting ferrite segregation, apparently held to the view that, in regions of high phosphorus concentration, austenite is actually lower in carbon content, the temperature of Ar_3 being thus raised. If this be a true picture of the behavior of the iron-carbon system, when cooling from the γ-zone and when phosphorus is segregated, then the opposite effect should be noticed in

hypereutectoid steels. For here, in the high phosphorus regions (the reasoning may be applied to other segregated elements, as well) the per cent of carbon in solution should again be lower, but the temperature of phase change, upon cooling, should be lowered, rather than raised. In other words, instead of finding cementite segregation around inclusions, we should expect to find the inclusions, in the majority of instances, in pearlite grains.

Mahin and Wilson (5) indirectly investigated this question by following an experimental method which had been developed for hypoeutectoid steels. Rods of various materials, ferrous and nonferrous, were driven into holes in a hypereutectoid steel, after which the piece was given a protracted treatment at high temperatures. Cementite segregation was observed after cooling. This is believed to be due to the inclusions thus artificially produced by the migration of traces of the foreign material into the steel.

Experimental

In the present investigation the question was approached from a different standpoint. All manufactured steels freeze over a considerable range of temperature, represented by the temperature zones enclosed between the liquidus and the solidus. This gives opportunity for the co-segregation of inclusions and dissolved impurities, due to the slope of the liquidus and solidus surfaces of the phase diagram, by the self purification of crystallizing austenite which has been held by Stead to account for the later separation of primary ferrite in the regions of nonmetallic inclusions. In order to avoid this action, high carbon steels have been prepared by cementation, using as starting materials four different specimens of nearly carbon-free iron: (1) Armco ingot iron, (2) Swedish charcoal iron, (3) Aston process wrought iron and (4) a wrought iron railway rail, contributed by the Louisville and New Albany Railway Co. and, presumably, dating back to fairly early railway history. The metallic phase of none of these irons was presumed to be pure iron but all were practically carbon-free and the composition and method of manufacture must be presumed to have allowed only a very limited freezing temperature range.

Specimens, in the form of square prisms, about one inch on a side, of each of these irons were cleaned, packed in a commercial carburizer and heated at 970 degrees Cent. (1780 degrees Fahr.). Approximately four hours were required to bring the specimens

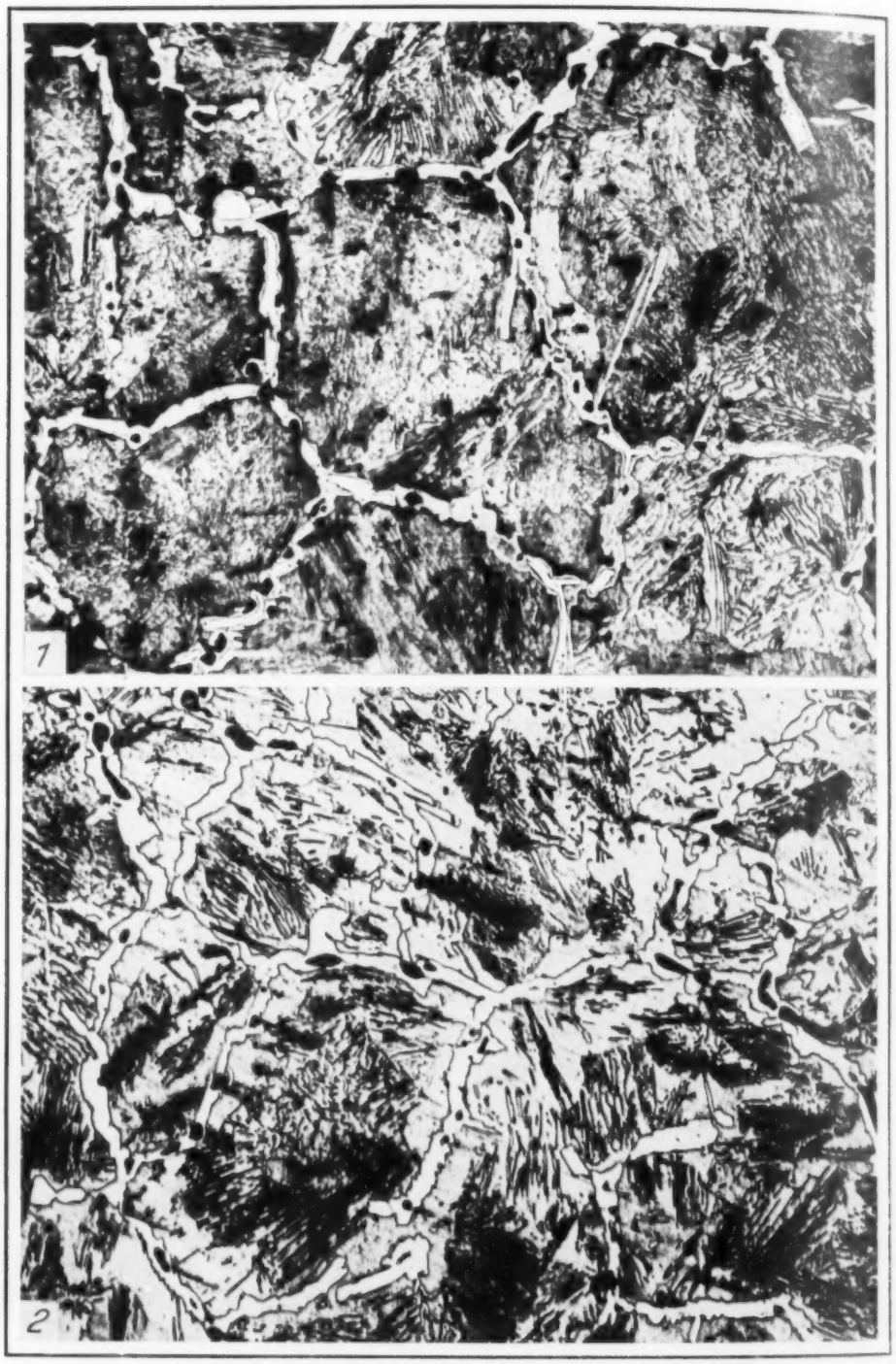

Fig. 1—Armco Ingot Iron. × 150.
Fig. 2—Armco Ingot Iron. × 250. Specimens were Carburized for 192 Hours at 970 Degrees Cent. (1780 Degrees Fahr.) and Cooled in the Carburizing Box.

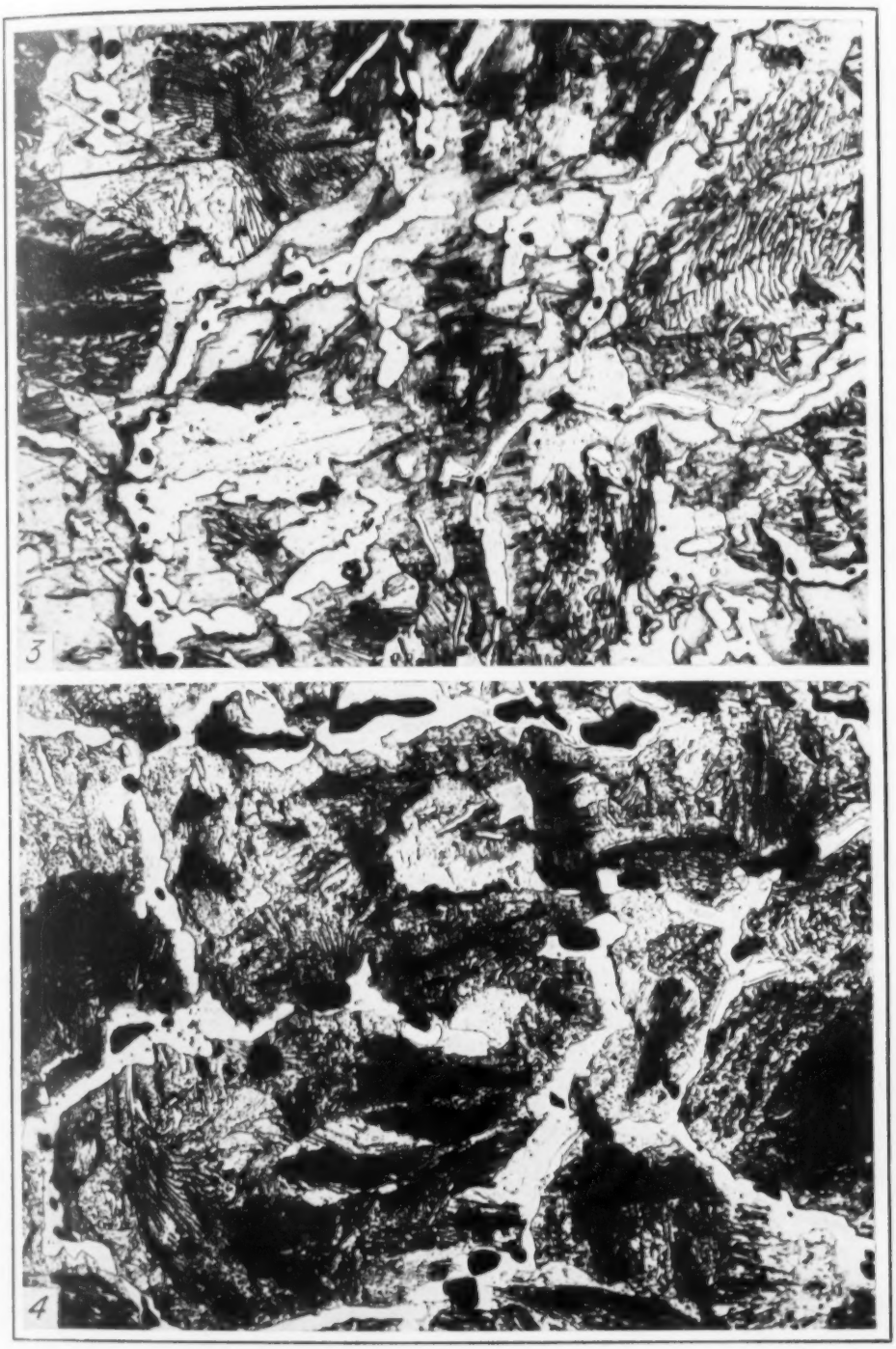

Fig. 3—Swedish Puddled Charcoal Iron. × 150.
Fig. 4—Aston Process Wrought Iron. × 150. Specimens were Carburized for 192 Hours at 970 Degrees Cent. (1780 Degrees Fahr.) and Cooled in the Carburizing Box.

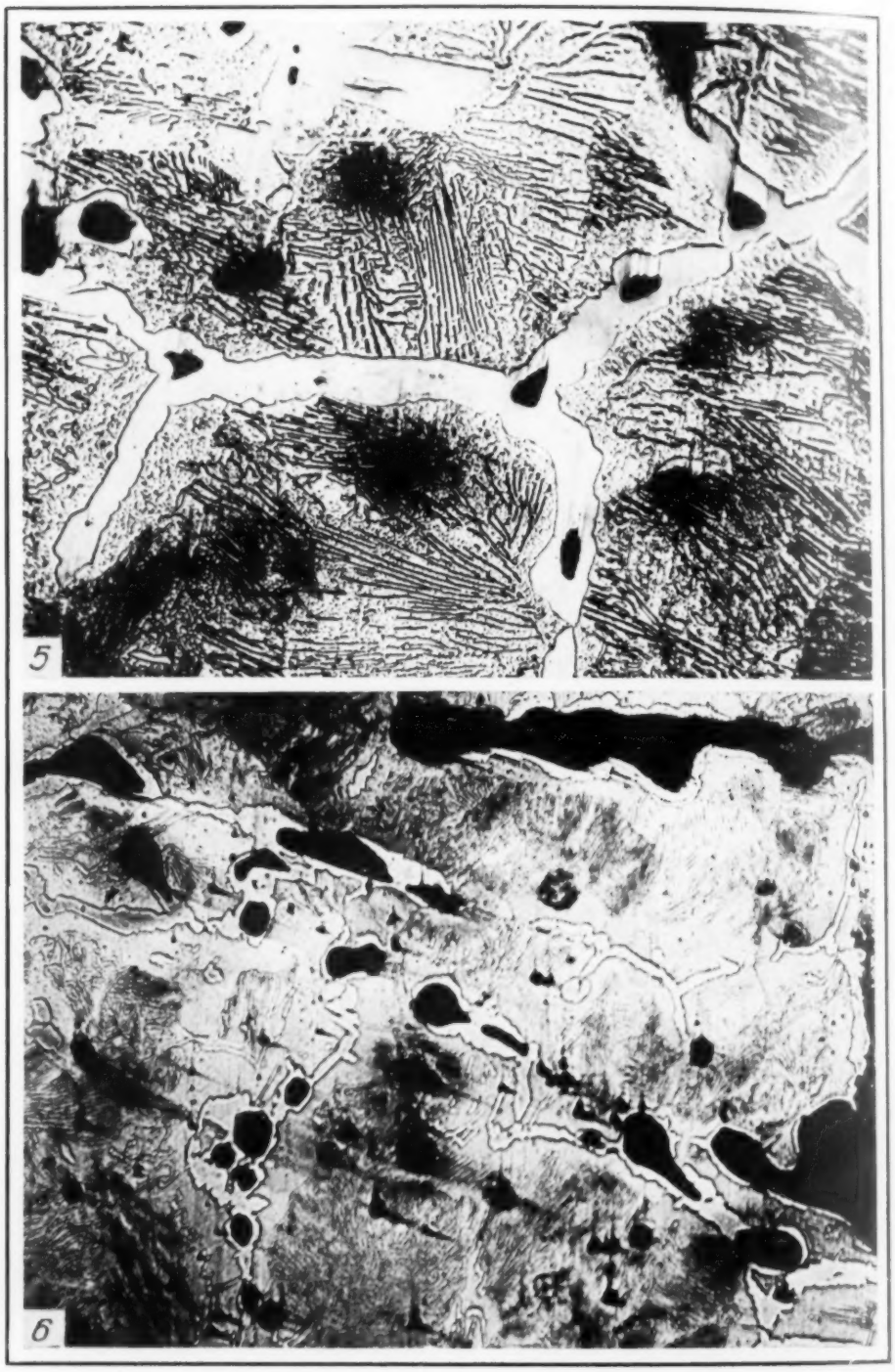

Fig. 5—Aston Process Wrought Iron. × 350.
Fig. 6—Puddled Rail. × 250. Specimens were Carburized for 192 Hours at 970 Degrees Cent. (1780 Degrees Fahr.) and Cooled in the Carburizing Box.

to temperature, which was then held for 192 hours. After cooling to below Ar_1 (requiring approximately three hours) the pieces were removed, cooled, sectioned and prepared for microscopic examination. The pieces were found to have been carburized to a depth of about three-eighths inch, with a considerable portion of the case decidedly hypereutectoid in character.

Similar results were noted in all four cases. Primary cementite showed a pronounced preference for the regions occupied by non-metallic inclusions and the pearlite grains were considerably divorced, thus yielding a heavy network of cementite, in which most of the inclusions were to be found. The photomicrographs of Figs. 1 to 6, inclusive, are typical illustrations of these structures.

Discussion

Although a chemical analysis of the materials used in the investigation has not been made, at least it may be said that Armco iron is known to be a highly purified commercial iron. Kenyon gives (9) the following statement as representing the average composition of this iron:

	Per Cent
Carbon	0.013
Manganese	0.017
Phosphorus	0.005
Sulphur	0.025
Silicon	trace
Total	0.060+

Such a metal would necessarily freeze within a temperature range of only a few degrees and the already small quantity of dissolved impurities would find little opportunity for segregation during the freezing process. Moreover, the long time period of heating for cementation, and the relatively high temperature employed, certainly provided ideal conditions for diffusion, so that dissolved impurities may safely be assumed to have become practically uniform in distribution, except as they may have been continuously contributed to the austenite by discrete particles of inclusions. Even as notoriously sluggish an element as phosphorus could no longer remain segregated (admitting a possibility of slight segregation in the original iron) as it has previously been shown (1) that a much shorter heating period than that here employed was sufficient to destroy phosphorus banding in an ordinary hypoeutectoid steel. That such impurities

were so contributed appears to be certain, from the results of the treatment.

Except for a probable lower degree of purity, the case of the Aston process iron would not be materially different, for this also has been blown in the converter to free it, so far as possible, from carbon and other impurities and the later addition of liquid slag should not materially affect the fundamental distinction in freezing characteristics between this and medium or high carbon steel. And the same opportunity has been provided in this, and in the other two specimens of puddled iron, for equalization of concentration during the cementation period. The cementation steels made from the two samples of puddled iron are similar in structure to the cemented Armco iron and Aston process iron.

Nonmetallic inclusions in these samples do not universally lie in the primary cementite network, but the great majority of them are so located. The tendency of the network to follow the locus of the inclusions is unmistakable. Inspection of the photomicrographs shows that, in a great many cases where an inclusion has been located near the obvious course of the network of cementite, the latter has curved aside to include it, and there is generally an enlargement in the network as it passes around the inclusion. This is only another way of saying that cementite precipitation has started in the zones about inclusions and that the network (cell, in three dimensions) has been developed by spread from these centers.

Divorcement

Attention has already been called to the excessive pearlite divorcement that has occurred in all of the specimens. Since all of the hypereutectoid zones were produced by pack carburization, it would be stated that all four materials were "abnormal", in the sense in which this term is now used. The authors have no desire to add to the confusion—if such it is—in connection with the attempt to understand the mechanism of what takes place during the carburizing and later cooling of abnormal steels. Probably nearly everyone accepts the view that oxygen in the steel is primarily responsible for abnormality and it is noted further that Grossmann has presented evidence (10) to indicate that it is not total oxygen, but only oxygen which has dissolved during the carburizing process, that is here significant and, even then, only in materials of low manganese content. It would appear that the oxygen so dissolved can hardly be chemically

free, but rather that of one of the oxides of carbon—probably carbon dioxide, formed from carbon monoxide in the surface layers of the steel by the reaction: $2CO \rightarrow CO_2 + C$.

In the course of discussions of the pearlite divorcement that takes place in the hypereutectoid zones of carburized abnormal steels, the term "diffusion" is often used to express the movement of pearlite-cementite to join the primary cementite in the network. This term appears not to be a fortunate one, in this connection, since it properly applies to the movement of elements or compounds in solution, in obedience to the force of osmotic pressure. The movement of cementite, during the progress of divorcement, is of a distinct phase already formed, analogous to the coalescence of particles dispersed in an emulsion. Coalescence is always the result of a difference between the surface tensions of the two phases composing the emulsion and it is always the particles of the phase of higher surface tension that coalesce to form masses of lower relative surface.

With this distinction in mind it seems obvious that "abnormality" is caused by some influence which widens the difference between the surface tensions of ferrite and cementite, by lowering that of ferrite or raising that of cementite, or by both effects. Whether it be oxygen, in some form, already contained in the steel, oxygen dissolved during carburization, or the presence or absence of excess manganese (as Grossmann believes) or of some other element or elements, that is responsible for the phenomenon of abnormality, it would appear (a) that this influence widens the difference in surface tension between ferrite and cementite and (b) that the elements which cause divorcement are segregated, either in the body of the austenite grains before the phase transformation begins or in the austenite grain boundaries where the cementite network will begin to form in the cooling steel. In the present paper, no answer to this question is suggested but it is believed that this aspect of the matter deserves further attention.

Harder and his co-workers (11), (12) supported the view that the presence of dissolved oxygen shifts the A_{cm} line of the iron-carbon diagram to the left, thus reducing the eutectoid carbon ratio. This would result in the formation of more primary cementite and less pearlite, but it is difficult to see any relation between this and the tendency of pearlite to divorce. Divorcement itself is apparently responsible for soft spots in the heat treated piece, the massive cementite returning to solution so slowly that low carbon austenite

in its vicinity transforms to troostite, instead of to martensite, upon quenching.

Conclusions

Nonmetallic inclusions affect the locus of the primary cementite network in hypereutectoid steels, through a mechanism similar to that operating in the precipitation of primary ferrite in hypoeutectoid steels. Dissolved impurities are segregated in the austenite zones which are in the immediate vicinity of the inclusions, and these segregated solutes cause the beginning of the primary phase formation before Ar_{cm} is reached for other parts of the solid solution. Probably this action is due to a diminished thermal hysteresis of the material about the inclusion, due to the presence of the dissolved impurity.

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11. O. E. Harder, L. J. Weber and T. E. Jerabek, "Studies on Normal and Abnormal Carburizing Steels", Transactions, American Society for Steel Treating, Vol. 13, 1928, p. 961.
12. O. E. Harder and W. S. Johnson, "Solubility of Carbon in Normal and Abnormal Steels", Transactions, American Society for Steel Treating, Vol. 15, 1929, p. 49.

DISCUSSION

Written Discussion: By H. B. Pulsifer, metallurgist, Ferry Cap and Set Screw Co., Cleveland.

In reading over the paper it seems to me that the authors may have been hasty in assuming that the black spots in their carbide networks are inclusions originally in the irons used.

One recalls that Professor Arnold, at that time of University College, Sheffield, as long ago as 1898 found that after ten cementations some two per cent of carbon could be induced into iron but that it came out largely as graphite in the cold metal.[1]

If the black spots in Mahin and Lee's paper are to be adjudged as inclusions, instead of graphite from the decomposition of cementite, they should show the characteristics of inclusions. The spots in question very much resemble the depressions left after rubbing down the soft surfaces of precipitated graphite. Fig. 6 of the paper indicates that a little longer heating, or cooling period, would have converted all of the massive cementite to graphite.

Oral Discussion

T. D. YENSEN:[2] I would like to ask a question which is somewhat apropos of the main topic, namely what the distinction is between a nonmetallic inclusion and a metallic inclusion.

E. G. MAHIN: I do not know that I can answer that question. I suppose the inference is that we do not need the term "nonmetallic."

At this moment I cannot say that I can think of any metallic inclusions. The term nonmetallic, however, has been used from the very beginning in such discussions, and I suppose we keep on using it because other people have used it before us.

T. D. YENSEN: I have discussed this matter with my colleagues, and I have not found anybody who has really a very satisfactory reply—maybe Mr. Ziegler has one now. It is with this as with so many other things—we start to use a term carelessly, and we just go on using it irrespective of whether it has any sense or not. After thinking about it, it seems that we could very well do away with "nonmetallic" and just talk about inclusions.

N. A. ZIEGLER:[3] There is no definite boundary line between structural constituents and inclusions in alloys. Both of them may be considered as members of binary or more complicated systems, and the difference is only in the relative solubility. For example, physically a cementitic particle in steel is just as much an inclusion as a particle of manganese sulphide, but it is not customary to refer to it as such. Both iron carbide and manganese sulphide form definite systems of alloys with iron. In regard to the former reference may be made to the well known iron-carbon diagram, and in regard to the latter, to the diagrams by Benedicks and Löfquist in their volume on the inclusions in iron and steel. The difference is that while iron carbide may be dissolved in iron to a considerable extent, the solubility of manganese sulphide is low.

[1] *The Metallographist*, Vol. 2, p. 64.

[2] Metallurgist, Westinghouse Electric and Manufacturing Company, East Pittsgurgh.

[3] Metallurgist, West Penn Power Company, Pittsburgh

The customary distinction is that by "inclusions" are meant particles visible under a microscope in an unetched metallic surface, while particles visible only after etching are referred to as "structural constituents."

As an example of metallic inclusions, lead in copper and its alloys may be considered. Since lead particles are visible in polished and un-etched surfaces of copper alloys, I think they should be classified as inclusions.

E. G. MAHIN: In case of the lead-copper alloy, I should not call these separated phases inclusions because they are essentially parts of a system. While we speak of nonmetallic inclusions, and the term may be entirely illogical and unnecessary, it is true just the same that the nonmetallic inclusions are usually unwanted things—not essential parts of the system. We cannot get rid of them but we do not care anything about them and we do not care to have them present. When we speak of cementite or any other essential phase of a metallic alloy that is a thing that we expect to be present, and by varying thermal treatment we alter the condition of these phases and bring about the production of desired properties in the alloy. So we should be very reluctant to speak of any distinctive phase in a metallic alloy as an inclusion. The nonmetallic inclusion may be essential from the standpoint that we cannot get rid of it but it is not anything that we have tried to produce or to have present. Possibly we should discontinue the term "nonmetallic"; as a matter of fact, a great many people have done so and I do myself frequently speak simply of "inclusions."

D. F. MCFARLAND:[4] It does not pay to quibble, but when a man says that the nonmetallic inclusion, as he calls it, is always unwanted, I wonder if that is true. I have understood that tungsten filaments have certain inclusions in them to perform a very definite office. I have understood that same thing is true of aluminum alloys, and certain other kinds of alloys, where nonmetallic substances are introduced for the definite purpose of restricting grain growth, so I do not believe that "unwanted" is exactly correct.

E. G. MAHIN: That is a question then for the committee on nomenclature.

W. J. DIEDERICHS:[5] I would like to ask a question of Professor Mahin. I wonder if he has studied the ferrite ring which forms around the temper-graphite spot in white iron, partially malleableized, and whether he would consider that the original temper graphite spot there acts as an inclusion to promote the separation of ferrite from the solid solution immediately surrounding that graphite spot. The question is briefly, "Does he consider that the temper graphite spot acts as a definite inclusion in that instance?"

E. G. MAHIN: Here is another case where the inclusion is perhaps not unwanted. I do not know that I have thought very much about that. I presume that I have had the same idea that others who have not studied the question particularly have had, that the ferrite surrounding the graphite spot has been due to the formation of graphite in the first place, that is to say that the formation of graphite has left uncombined iron surrounding it. I do not know that that is correct. Just at present, I cannot see that there would be any relation between this graphite inclusion and the austenite which has existed at higher temperature,

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[4]Head, department of Metallurgy, Pennsylvania State College, State College, Pa.

[5]Metallurgist, Autocar Company, Ardmore, Pa.

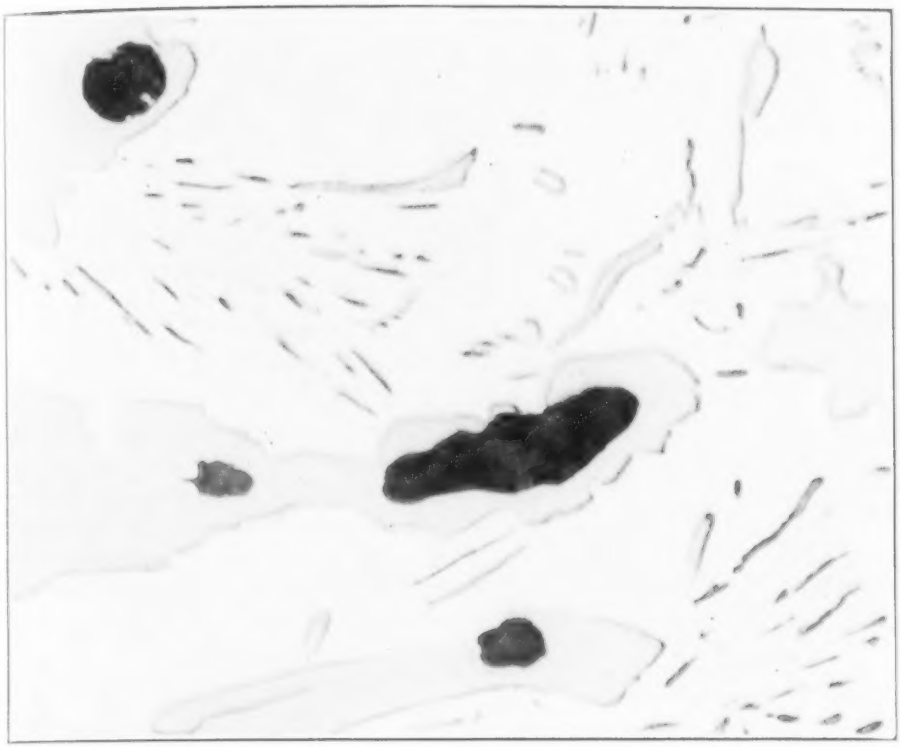


Fig. 7. Armco Ingot Iron. Specimen Carburized for 192 Hours at 970 Degrees Cent. (1780 Degrees Fahr.) and Cooled in the Carburizing Box. Specimen Polished but Not Etched. × 650.

analogous to that existing between the nonmetallic inclusion and austenite. I am not prepared to definitely answer that question.

The remarks of Mr. Pulsifer were received some weeks after the date of reading the paper. On account of the unusually dark appearance of the inclusions, the suspicion that at least some of them might be graphitic carbon was apparently justified. This dark appearance was enhanced by the photographic process, since we were primarily concerned with etching and photographing in such a manner as to bring out, as strikingly as possible, the microstructure of the metallic phases. With Mr. Pulsifer's criticism in mind, the specimens have been very carefully re-examined, at both low and high magnifications. In no instance was a graphite inclusion (or a pit formed by rubbing one out) found, all of the inclusions, whether rounded or elongated, being of dark gray glassy appearance or having the mottled character of slag of two phases.

It might be remarked also that if any graphite had been formed by graphitization of massive cementite, such inclusions should be bordered by ferrite. No such structure was detected in any of the specimens.

One additional photomicrograph has been made at 650 diameters, the piece (of carburized Armco iron) having been repolished and photographed without previous etching. On account of the massive character of cementite, relief polishing has revealed the structure fairly well. This is shown in Fig. 7.

DIRECTIONAL PROPERTIES IN ROLLED AND ANNEALED LOW CARBON STEEL

By Arthur Phillips and H. H. Dunkle

Abstract

This paper includes simple mechanical tests on hot- and cold-rolled low carbon steel and cupping and mechanical tests on strip annealed at four temperatures ranging from 680 to 745 degrees Cent. Tests were made on specimens cut at 0, 22.5, 45, 67.5 and 90 degrees to the rolling direction. In the material given heavy reductions prior to the annealing treatment, a maximum in tensile strength and a minimum in elongation values were found at about 45 degrees. The height of the ears on drawn cups increased with the degree of cold rolling. The strip rolled 40 per cent formed ears 45 degrees to the rolling direction. From the strips rolled 50, 60 and 69 per cent ears were located in all cases at the 0 and 90-degree positions.

THIS paper is one of a series devoted to the consideration of directional properties in cold-worked and annealed metals. With the exception of the results reported at this time our studies have been mainly restricted to copper and copper alloys[1, 2]. The general object of the program is to furnish a desirable compilation of mechanical properties showing the significance of fiber effects produced by a diversified schedule of working and heat treatment which will in addition serve as a statistical background for a proper evaluation of X-ray studies on fibered materials. Although fibering has been studied by X-ray methods by a number of workers, no serious attempt has been made to correlate the results with the behavior of the metal under deformational treatment.

In general, two distinct types of fiber may be produced in wrought steel. One form, commonly known as mechanical fibering,

[1]Arthur Phillips and E. S. Bunn, "Directional Properties in Cold-Rolled and Annealed Copper," *Transactions*, American Institute of Mining and Metallurgical Engineers, 1931, Vol. 93, Institute of Metals Division, 353.

[2]Arthur Phillips and C. H. Samans, "Directional Properties in Cold-Rolled and Annealed Commercial Bronze," *Transactions*, American Institute of Mining and Metallurgical Engineers, 1933, Vol. 104, Institute of Metals Division, 171.

A paper presented before the Sixteenth Annual Convention of the Society held in New York City, October 1 to 5, 1934. Of the authors, Arthur Phillips is professor of metallurgy and H. H. Dunkle is graduate student, Department of Metallurgy, Yale University, New Haven, Conn. Manuscript received June 25, 1934.

results from the extension of segregated constituents such as dendrites, slag, gas holes, carbides and sulphides. Fibering of this kind is usually evidenced by the presence of stringers or bands extended in the direction of working and in many cases may be detected by the character of the fracture. The second type, essentially a crystallographic fiber, is produced by working. By cold-working a metal the grains are elongated and continued reduction usually results in a condition of parallelism in the structure. Furthermore, under certain conditions of working there is a pronounced tendency for the crystalline fragments to become oriented in certain directions with respect to the axis of extension. Even after annealing following cold-working preferred orientations may be present.

Howe[3] studied the causes and practical significance of fiber in

Table I
Composition and Rolling Treatment

Coil No.	Approx. Finishing Temp. Degrees Cent.	Hot-rolled to Inch	Analysis of Composite Sample Composed of Drillings Taken Across 7-Inch Strip			
			Carbon	Manganese	Phosphorus	Sulphur
1	730	0.053	0.07	0.37	0.008	0.032
2	775	0.065	0.08	0.40	0.006	0.032
3	800	0.079	0.08	0.40	0.006	0.034
4	830	0.103	0.07	0.36	0.007	0.030

steel and wrought iron. His assembly of data, covering some 1700 tests, showed that wrought iron and puddled steel sheets (materials of plastic origin) were weaker transversely than longitudinally. The transverse deficit ranged from 5-29 per cent. The longitudinal test pieces also yielded the better elongation values. In the steels of molten origin the differences in tensile properties were less pronounced. In general, however, the longitudinal tests were distinctly superior in strength and elongation values. It should be noted that the data compiled by Howe were obtained from many sources and, furthermore, most of the tests were made prior to 1890. The testing results collected by Howe without doubt show the effect on properties of mechanical fibering resulting from the relatively large amount of slag and sonims present in the material of those days but are of little value in studying the nature and significance of crystallographic fibering.

The present paper is concerned essentially with simple mechanical tests on hot-and cold-worked low carbon steel, and cupping and

[3]H. M. Howe, Metallography of Cast Iron and Steel," McGraw-Hill Book Co., 1916.

mechanical tests on the annealed steel. No attempt has been made to include a study of the age-hardening characteristics of the steel. In general, all tests were made very promptly after the material was received from the mill in order to avoid complications which might be due to aging.

Preparation and Testing of Hot-Rolled Material

Four ingots were selected from an open-hearth heat which had a ladle analysis as follows:

	Per Cent
Carbon	0.09
Manganese	0.40
Phosphorus	0.008
Sulphur	0.028

One slab, taken from near the top of each ingot, was set aside for the hot-rolling treatment. The four slabs were then hot-rolled into 7-inch strips differing only in gage as shown in Table I. This resulted in four coils averaging about 550 pounds.

Tensile strength and elongation tests were made on specimens representing five directions in the hot-rolled strip, viz., 0, 22.5, 45, 67.5, and 90 degrees to the axis of rolling. The results of the tests are summarized in Table II and are shown graphically in Fig. 1. The maximum variation in tensile strength of the several strips was found in the material rolled to 0.053 inch, i.e., the steel having the greatest reduction. In this case, the transverse strength was found to be some 6 per cent greater than that of the longitudinal specimens. The corresponding difference in the material rolled to 0.065 inch was only about 3 per cent. No attempt was made to determine by cupping tests whether or not directional tendencies of the order noted in these tests have any significance with regard to the formation of ears on drawn cups. The hot-rolling program was planned to furnish a reasonably uniform series of strip, differing only in thickness, which would be best adapted for the investigation of cold-reduction effects and subsequent annealing treatments.

Tests on Cold-Rolled Steel

Approximately 150 pounds of strip were cut from the center of each of the 550-pound coils of hot-rolled steel. The material thus located had cooled slowly and relatively uniformly from the finishing temperature. After thorough pickling, the strips were

cold-rolled to a common gage, 0.032 inch. This treatment resulted in four strips of identical dimensions but differing markedly in the total reduction by cold-rolling. The reductions were 40, 50, 60, and 69 per cent (see Table II).

Tensile tests were made on these strips in the five directions with respect to the rolling axis. The results are shown in Table III and

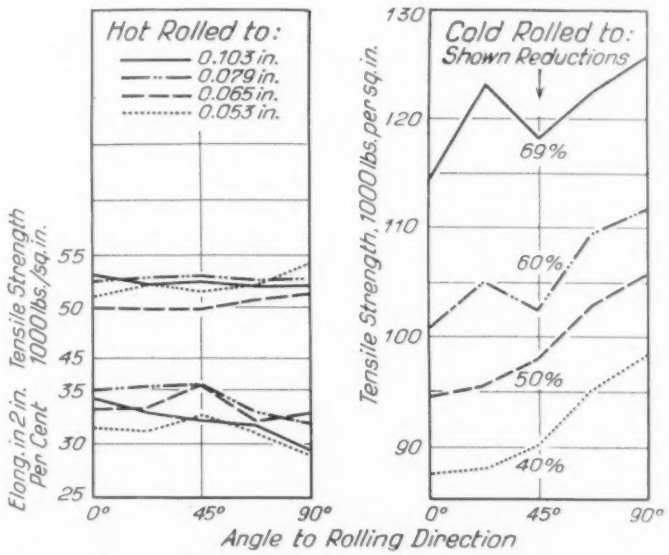


Fig. 1—Tests on Hot-and Cold-Rolled Material.

Fig. 1. In general, it may be stated that, in all cases, the transverse strength is the greatest; furthermore, the strength values increase progressively from the 0 to the 90-degree positions, with the exception that a peak is evident in the 22.5-degree direction in the strips reduced 60 and 69 per cent. The transverse strength is some 10 to 12 per cent greater than the longitudinal. Moreover, the percentage difference does not seem to be a function of the degree of cold-working within the range of reductions studied.

Our results on copper and 90-10 brass have shown a steady increase of strength from the 0 to the 90-degree directions. The transverse strength of severely rolled (90 per cent reduction) tough-pitch copper is only approximately 3 per cent greater than in the longitudinal direction; in 90-10 brass, reduced 95 per cent, the transverse superiority was found to be about 17 per cent.

Tests on Annealed Steel

After testing the cold-rolled strips, each of the four coils was

Table II
Properties of Hot-Rolled Steel

Hot-Rolled to Inch	Rockwell B-Scale	Tensile Strength Pounds per Square Inch*					Elongation in 2 Inches Per Cent*				
		0°	22.5°	45°	67.5°	90°	0°	22.5°	45°	67.5°	90°
0.103	60	53,000	52,250	52,250	52,200	52,000	34	32.8	31.8	31.5	29.3
0.079	59	52,300	52,700	52,750	52,500	52,800	35	35.5	35.5	33	32
0.065	55	49,900	49,620	49,680	50,730	51,280	33.5	33.5	35.5	32	33
0.053	59	51,080	52,250	51,520	51,900	54,160	31.5	31	33	31	29

*Average value of three tests.

cut into four parts. One length from each coil was annealed at 680 degrees Cent., a second at 705 degrees Cent., a third at 720 degrees Cent. and the fourth at 745 degrees Cent. In each heat treatment the specimens were brought with the furnace to temperature, heated for two hours, after which the metal was furnace-cooled to ordinary temperatures. All heating and cooling was carried on in an atmosphere of producer gas.

The results of the hardness, tensile and Ericksen tests made on the sixteen sets of material are given in Table IV. In addition to these measurements, cupping tests were made and the location and heights of the ears are also included in the Table IV. The figures shown represent the average heights of the ears above the adjacent valley. Tensile and elongation values are shown graphically in Fig. 2.

With respect to the tensile properties of the strip annealed after the lowest reduction (40 per cent) it is interesting to note that, although the strength values have markedly decreased as would be expected, the curves in general are similar to those for the cold-worked materials, in that there is a more or less progressive increase in strength from the 0 to 90-degree positions. In other words, the type of fiber induced by cold-working has persisted after the annealing treatments. The anneals following the 50 and 60 per cent reductions tend to shift the maximum tensile strength value from the 90 to the 67.5 and 45-degree directions. In the case of the anneals subsequent to the 69 per cent reduction, however, there is a decided maximum at the 45-degree position. The elongation values, on the other hand, for the 69 per cent reduction reach a minimum value in the neighborhood of the 45-degree direction. It does not seem wise to attempt to correlate the minor differences in properties with the heat treatments for a given reduction. It is quite probable that the slight irregularities in treatment almost unavoidable in a mill

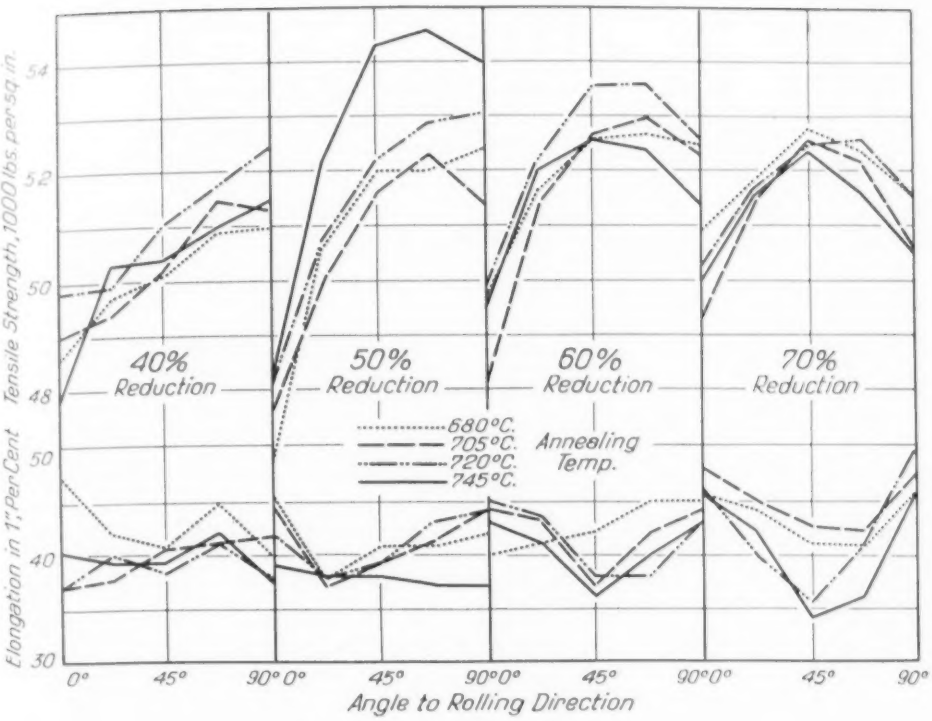


Fig. 2—Tests on Annealed Steels.

program, may be responsible for greater differences within a group than may be properly attributed to the planned heat treatments within the annealing range considered.

Turning now to the cupping tests, which are probably a more significant index of directional effects, it is evident that the height of the ears increases with the degree of cold-rolling prior to the annealing treatment. Generally speaking this condition has been noted by the authors in copper and copper-base alloys. Disregarding for the moment the material rolled 40 per cent, it is interesting to note that the relative ear heights for the 50, 60 and 69 per cent reductions are 1, 2 and 3. For these reductions, the ears are located in the 0-degree and 90-degree directions with respect to the rolling axis, locations commonly observed for the tips found on drawn steel articles. In the case of the material rolled 40 per cent, however, the ears were found without exception at 45 degrees to the rolling direction. As far as the writers are aware, ears are seldom, if ever, noted in the 45-degree position in commercially drawn steel articles.

It is interesting to note that the greatest earing tendencies, from the standpoint of dimensions, were found in the material showing a

Table III

Properties of Cold-Rolled Steel

Series	Reduction Actual-In.	Reduction %	Rockwell B-Scale	Tensile Strength Pounds Per Square Inch 0°	22.5°	45°	67.5°	90°	Elongation in 2 Inches Per Cent				
1	0.053 to 0.032	40	84	87,350	88,800	90,500	95,650	98,550	2	2.5	3	3	2.5
2	0.065 to 0.032	50	87	94,500	95,850	97,900	103,000	106,000	1.5	1.5	2	1.5	1.5
3	0.079 to 0.032	60	91	101,000	105,100	102,500	109,000	111,350	2	1.75	2	2	2
4	0.103 to 0.032	69	94	114,700	123,450	118,500	122,500	125,700	1.75	1.5	1.5	1.5	2

maximum strength value and a minimum elongation value in the 45-degree direction. In all such cases the ears were found at the 0 and 90-degree positions with respect to the axis of rolling. On the other hand, tough-pitch lake copper, annealed at high temperatures following heavy reductions, has a decided elongation maximum in the 45-degree direction but also develops ears at the 0 and 90-degree locations. Although the explanation for this apparent anomaly is not clear we must at least bear in mind that we are dealing with materials having different lattice characteristics, namely, that the alpha iron is body-centered cubic while the copper is face-centered cubic.

Table IV

Properties of Annealed Steel

Reduction Per Cent	Temperature of Anneal °C.	Tensile Strength Pounds Per Square Inch 0°	22.5°	45°	67.5°	90°	Elongation in 1 Inch Per Cent 0°	22.5°	45°	67.5°	90°	Rockwell B-Scale	Ericksen	Height Ears, In.†
40	680	48,600	49,700	50,100	50,900	51,000	47	42	41	44.5	40	48	10.9	0.018*
40	705	49,000	49,400	50,200	51,500	51,300	37	38	41	41.5	42	45	10.8	0.017*
40	720	49,800	49,900	51,100	51,750	52,500	37	40	38.5	41.5	38	46	10.7	0.015*
40	745	47,800	50,300	50,400	51,000	51,500	40.5	39	39	42	37.5	46	10.7	0.015*
50	680	46,500	50,600	52,000	52,000	52,450	46	38	41	41	42	48	11.0	0.012
50	705	47,500	49,900	51,600	52,300	51,450	45	37	39	41.5	44	49	10.7	0.013
50	720	47,900	50,700	52,200	52,900	53,100	42	38	39	43	44	48	10.6	0.013
50	745	48,000	52,200	54,300	54,600	54,000	39	38	38	37	37	48	10.6	0.014
60	680	49,700	51,700	52,600	52,700	52,000	40	41	42	45	45	52	10.7	0.024
60	705	48,200	51,400	52,700	53,000	52,300	44	43	37	42	44	50	10.8	0.022
60	720	49,800	52,200	53,500	53,600	52,600	45	43.5	38	38	43	51	10.7	0.023
60	745	49,500	52,000	52,600	52,400	51,400	43	41	36	40	43	50	11.0	0.022
69	680	50,900	51,800	52,800	52,400	51,600	45.5	44	41	41	46	52	10.6	0.037
69	705	49,300	51,400	52,600	52,200	50,700	48	45	42.5	42	47	52	10.8	0.035
69	720	50,200	51,600	52,500	52,600	51,600	46	40	35.5	41	49.5	52	10.6	0.038
69	745	50,000	51,600	52,400	51,600	50,500	45.5	42	34	36	46	50	11.0	0.039

†Average of measurements on 10 cups.

*Located 45° to direction of rolling.

Referring again to the curves of Fig. 2 it is evident that the material showing the maximum differences in tensile properties within a given strip does not necessarily develop the highest ears. For example, the steel rolled 40 per cent prior to the annealing treatment gave the maximum directional variations but developed ears only one third as high as those formed from the strip rolled 69 per cent.

The writers are indebted to the Stanley Works for the material used in their tests. They are glad to acknowledge the interest and assistance of R. L. Baldwin, chief chemist of the American Tube and Stamping Plant of the Stanley Works.

DISCUSSION

Written Discussion: By Robert F. Mehl, Director and M. Gensamer, Metals Research Laboratory, Carnegie Institute of Technology, Pittsburgh.

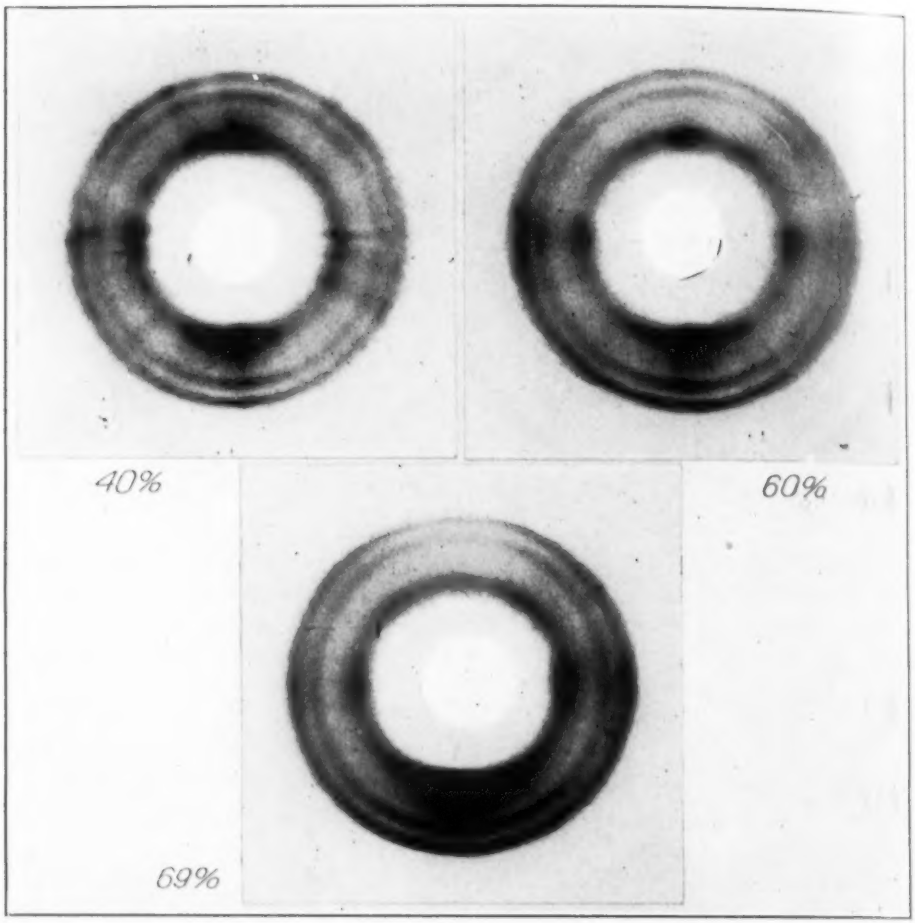
After so many thousands of tons of steel have been hot- and cold-rolled it is indeed pleasant to have Professor Phillips and Mr. Dunkle present this work on direction properties in steel sheet for it is, so far as we know, the first paper on the subject. There are so many interesting and important scientific problems in the manufacture and use of steel sheet that the lack of papers in this field is more than surprising.

Perhaps the most surprising thing about the results reported is the small difference in directional properties between the 40 per cent cold-rolled and the 69 per cent cold-rolled sheet shown in Fig. 1, as pointed out by the authors. One is inclined to ascribe these directional properties to the effect of the preferred orientations present in the sheet, and if there is, as other investigators have found, a gradual increase in the degree of preferred orientation as rolling progresses, the results of the authors are difficult to understand.

It happens that a study of the development of the preferred orientations produced by cold rolling in steel of a composition similar to that reported here is now being performed in our laboratory. The most complete work on this subject thusfar reported is that of Kurdjumow and Sachs,[1] who prepared a full pole figure for the preferred orientations in a sheet cold-rolled 98.5 per cent. These data showed that the complicated preferred orientations can be expressed as a combination of three simple orientations: (1) (1 0 0) parallel to rolling plane, [0 1 1] parallel to rolling direction; (2) (1 1 2) parallel to rolling plane, [1 1 2] parallel to rolling direction. Of these three orientations, numbers (1) and (2) are present in nearly equal amounts; number (3) is only a slight amount.

The work which we have done on the same subject has confirmed the work of Kurdjumow and Sachs. We have included in our studies samples cold-rolled from 5 per cent upwards and full pole figures have been prepared on a number of samples for both the inside and outside of the sheet. We have found that the *type* of preferred orientation just described is developed at the beginning of

[1] G. Kurdjumow and G. Sachs, "Walz- und Rekristallisationstextur von Eisenblech," *Zeitschrift für Physik*, Vol. 62, 1930, p. 592-599.



Photograms of Cold-rolled Material Rolled with the Percentage of Reduction Indicated. Photograms Taken with Plane of Sheet Perpendicular to X-ray Beam.

the process, though our photograms are insufficiently developed to permit an analysis in terms of pole figures below 40 per cent cold rolling. Between 40 per cent and 99 per cent there is a progressive sharpening of the diffraction nodes, but this may be partly caused by increasing fragmentation rather than by an increasing degree of preferred orientations. There is a slight difference in the distribution of the pole areas with increasing cold rolling and also a difference between the inside and outside of the sheet, but at no point between 40 per cent and 99 per cent is there any radical change in the *type* of preferred orientation. This is contrary to previous statements[3] based on single photograms. It must be stated that single photograms are of little use in evaluating types of preferred orientation or degrees of preferred orientations. The use of "four-point" and "six-point" photograms in characterizing two *types* of preferred orientations is especially to be condemned, for a very slight shift in orientation, either of the sample studied in relation to the direction of the

[3]W. A. Sisson, "Cold Rolling Low Carbon Sheet Steel," *Metals and Alloys*, Vol. 4, Dec. 1933, p. 193-198.

X-ray beam, or of the preferred orientations within the sheet will suffice to change the number of nodes on the diffraction ring. In such cases there may be a wholly unimportant change in preferred orientation.

Professor Phillips and Mr. Dunkle kindly supplied us with some samples of their cold-rolled material, namely those cold-rolled 40, 60 and 69 per cent. Photograms taken with the plane of the sheet perpendicular to the X-ray beam are shown in the accompanying Fig. We see that there is little change in the sharpness of the nodes as the degree of cold rolling increases. This is consistent with the results of the measurements of tensile strength, shown in Fig. 1, which shows little change in the directional effect. However, our samples show a more gradual development of the preferred orientation; and it is accordingly possible that our samples would show a progressive development in directional properties in this range. These differences are probably caused by differences in type of material, grain-size, and rolling technique, and may possibly be the result of the author's having rolled all their specimens from a variety of thicknesses to a common thickness, whereas our samples were rolled from a common thickness to a variety of thicknesses.

It is not possible to study preferred orientations in steel sheet below approximately 40 per cent cold rolling (at least in our materials), for the grain-size is sufficiently large to give a spotty film, without clearly defined nodes. This is a severe limitation in the X-ray method, though it is possible that special techniques might be developed to give better results.

It is surprising that the sound figure method developed by Tammann[3] some years ago has never been improved by the use of modern sonic and other driving mechanisms. This method, we believe, would be much more useful on low degrees of preferred orientation than the X-ray method, and many important industrial problems concern low rather than high degrees of preferred orientation.

N. P. Goss:[4] I understand that the diffraction patterns which Dr. Gensemer has shown were made with the X-ray beam normal to the plane of rolling.

I find that in order to get the maximum information concerning the distribution of the crystal fragments, the specimen must be set at various angles to the X-ray beam in order to study the distribution of some important crystallographic direction and to determine the position at which the grain fragments are oriented most strongly.

In the case of cold-rolled material I find that the limiting angle of preferred orientation for the 1 1 0 plane is about 20 degrees beyond which there is no orientation, the limiting angle varies depending upon the roll diameter and the amount of back tension used. Another thing which I have found is that the most intense fibre structure is developed in a transverse direction; that is, when the X-ray beam is transmitted parallel to the surface and at right angles to the direction of rolling. When X-rayed in this direction that even a small amount of cold work, or hot work, devolpes a very intense fibre structure. It can be seen therefore that not only is it important to study the distribution of

[3]G. Tammann and E. Schröder, *Zeitschrift f. Metallkunde*, Vol. 16, 1924, p. 202; G. Tammann and W. Riedelsberger, *Zeitschrift f. Metallkunde*, Vol. 18, 1926, p. 105, 149.

[4]Physicist in charge of metals research, Cold Metal Process Co., Youngstown, Ohio.

the grain fragments about some important direction but it is also of the greatest importance to determine the direction in which the fibre structure is most intense.

Authors' Closure

The discussions by Messrs. Mehl and Gensamer and Mr. Goss deal mainly with the X-ray diffraction characteristics of the cold-rolled steel. The authors, previous to the presentation of the paper, had made a short study of the fiber conditions of the several specimens of rolled material and their results lead to the conclusions expressed by Messrs. Mehl and Gensamer.

We are inclined to believe that the X-ray approach is likely to prove disappointing in the study of preferred orientations in metals slightly or moderately cold-worked. For this reason, we are pleased to learn that Dr. Mehl is considering other methods of attack. As he suggests, the sound figure method developed by Tammann may prove useful for determining the degree of preferred orientation in the earlier stages of reduction. Generally speaking, the most satisfactory fiber studies to date have been concerned with metals cold-worked most drastically and for that reason not comparable to the reductions included in mill schedules.

MALLEABLE CASTINGS

By Albert Sauveur and Harry L. Anthony

Abstract

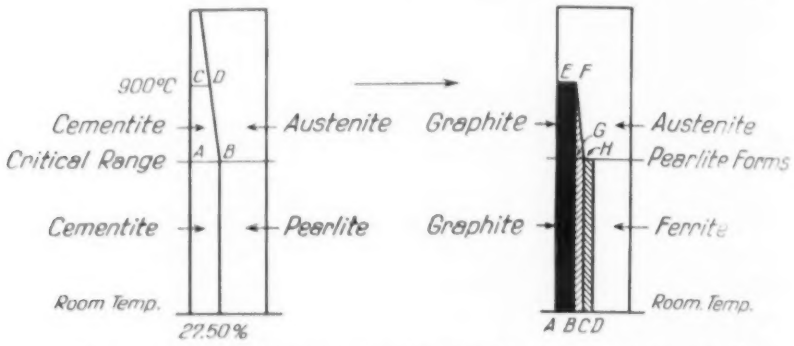
This paper includes an attempt at explaining the mechanism of the formation of graphite in the malleableizing process. Free cementite only graphitizes above the thermal critical range. The carbon present in the austenitic matrix above that range must be thrown out of solid solution through the formation of pearlite before graphitizing can take place. This process moreover causes the spheroidizing of the pearlite-cementite as a necessary step preceding its graphitizing. Through suitable annealing treatments it is possible to produce castings with different matrices and therefore possessing widely different physical properties. These matrices may be ferrrtic, pearlitic or sorbitic in character or they may be in the spheroidized condition when tensile strengths in excess of 100,000 pounds per square inch are readily obtained conbined with an elongation of some 7 per cent. The annealing treatment needs not exceed 40 hours.

THE art of imparting a certain amount of ductility and toughness to castings made originally of white cast iron and, therefore, extremely hard and brittle is a very old one. As practiced in the United States, the operation consists of a long annealing of the castings at a temperature well above their thermal critical ranges followed by very slow cooling. The articles now known as "black heart" castings are made up of many rounded particles of graphite (temper carbon) embedded in a matrix of ferrite. Their tensile strength may reach some 55,000 pounds per square inch and their elongation some 15 to 18 per cent in two inches. The specifications of the American Society for Testing Materials call for 50,000 pounds per square inch and 10 per cent elongation for one grade and 53,000 pounds per square inch and 18 per cent elongation for a better grade.

There seems to have been little desire, at least until quite recently, on the part of manufacturers to modify the matrices of malleable

A paper presented before the Sixteenth Annual Convention of the Society held in New York City, October 1 to 5, 1934. Of the authors, Dr. Sauveur is Gordon McKay Professor of Metallurgy and Metallography, Harvard University, and Harry L. Anthony is candidate for the degree of Doctor of Science in Metallurgy, Harvard University, Cambridge, Mass. Manuscript received May 30, 1934.

castings and thereby to produce different physical properties, notably greater strength. Makers of grey castings, on the contrary, are not satisfied with castings made up of graphite and ferrite. Through suitable control of composition, size of castings, rate of cooling, etc., they produce at will castings having hypoeutectoid, eutectoid or hyper-



Figs. 1 and 2—Diagrams Showing the Heating of 2.50 Per Cent Carbon White Cast Iron Having 72.50 Per Cent Pearlite and 27.50 Per Cent Cementite and the Structural Changes Which Take Place.

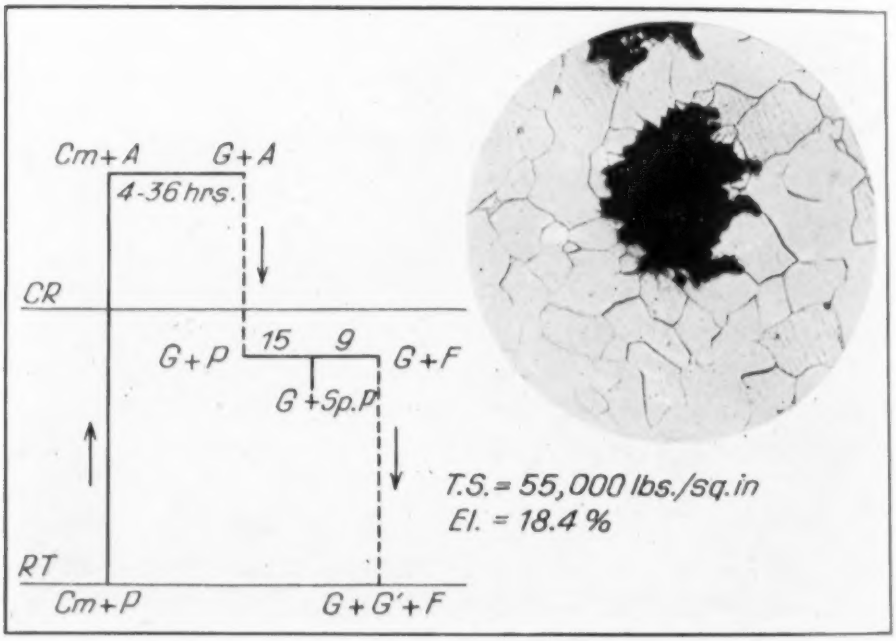


Fig 3

eutectoid matrices, pearlitic or sorbitic, and, consequently, varying much in tensile strength, hardness and ductility. These considerations led us to investigate the practical possibility of producing malleable castings in which the temper carbon may be embedded into sorbitic, pearlitic or spheroidized matrices. It should be borne in mind that

for like matrices malleable castings should be superior in properties to grey castings, because of the rounded form assumed by temper carbon compared with the well known occurrence of graphite in grey cast iron. It should be possible in malleable castings to combine great strength with suitable ductility to a much greater extent than in grey castings. The materials used in these experiments were obtained from various sources but the analysis of all brands were included within the following limits:

	Per Cent
Carbon	2.20-2.50
Silicon	0.85-1.00
Manganese	0.26-0.37
Sulphur	0.03-0.10

Mechanism of Graphitizing

Let us consider briefly the mechanism of the graphitizing of the carbide Fe_3C during the malleableizing process. Assuming the white casting to contain 2.50 per cent carbon, it will consist of approximately 72.50 per cent pearlite and 27.50 per cent free cementite.

On heating this casting through its $Ac_{3.2.1}$ point (Figs. 1 and 2), the pearlite is converted into austenite. At higher temperatures, some of this free cementite goes into solution in the austenite. At some 900 degrees Cent. (1650 degrees Fahr.), which may be the annealing temperature for malleableizing, the casting contains CD per cent of free cementite, possibly some 20 per cent. Given time, this cementite will graphitize; that is, will be converted into small particles of temper carbon, while the ferrite set free by this dissociation will diffuse into the surrounding austenite. The proportion of graphite at this stage may be represented by the distance EF in Fig. 2. As the iron now cools slowly to the critical range, some cementite is liberated from the austenite and it graphitizes, as indicated by the triangle F G H. Below the critical range, given the necessary time, the cementite of the pearlite spheroidizes and then graphitizes. Finally, after complete graphitizing, the casting will contain AD per cent of temper carbon, of which AB was formed by long annealing at 900 degrees Cent. (1650 degrees Fahr.), BC per cent by slow cooling from that temperature to the critical temperature and CD through the graphitizing, below the range, of the cementite present in pearlite. These distances are merely indicative of the mechanism of the dissociation of cementite during the various stages of the operations and are not to be taken as representing actual percentages.

Ferritic Matrix

Complete graphitizing resulting in a ferritic matrix is generally obtained by long annealing above the critical range followed by very slow cooling through and below that range, which may include the maintenance of a constant temperature below the range for a con-

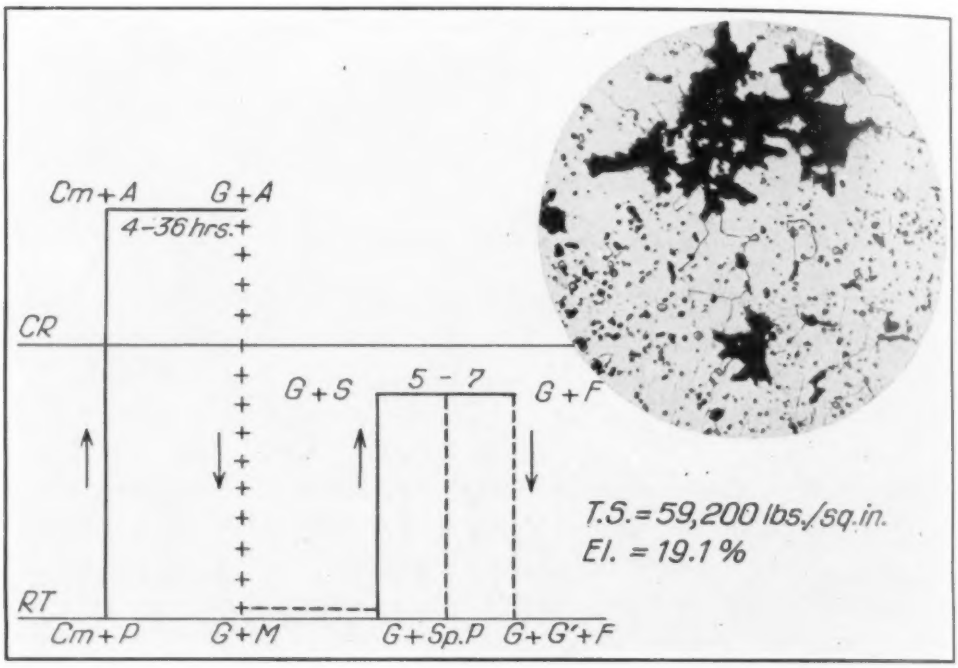


Fig. 4

siderable time. This has been depicted in Fig. 3. The time of annealing above the range may vary between 4 and 36 hours, being the shorter the higher the temperature. Relatively low temperatures; that is, not exceeding 950 degrees Cent. (1740 degrees Fahr.), are generally used.

In this and following diagrams, RT stands for room temperature, CR for critical range, Cm for cementite, A for austenite, G for graphite formed about the range, G′ for graphite formed below the range, Sp P for spheroidized pearlite and F for ferrite. The cementite present in pearlite must necessarily spheroidize before it graphitizes. In Fig. 3, it is intended to indicate that it requires about 15 hours to spheroidize the pearlite and 9 hours more to graphitize it.

The malleable casting obtained in this way had a tensile strength of about 55,000 pounds and an elongation in two inches of 18.4

per cent. Malleable iron with a ferritic matrix may also be obtained by quenching from the annealing temperature followed by reheating to some 650 degrees Cent. (1200 degrees Fahr.), thereby spheroidizing and graphitizing the sorbite formed at that temperature (Fig. 4). The tensile strength was 59,200 pounds per square inch and the

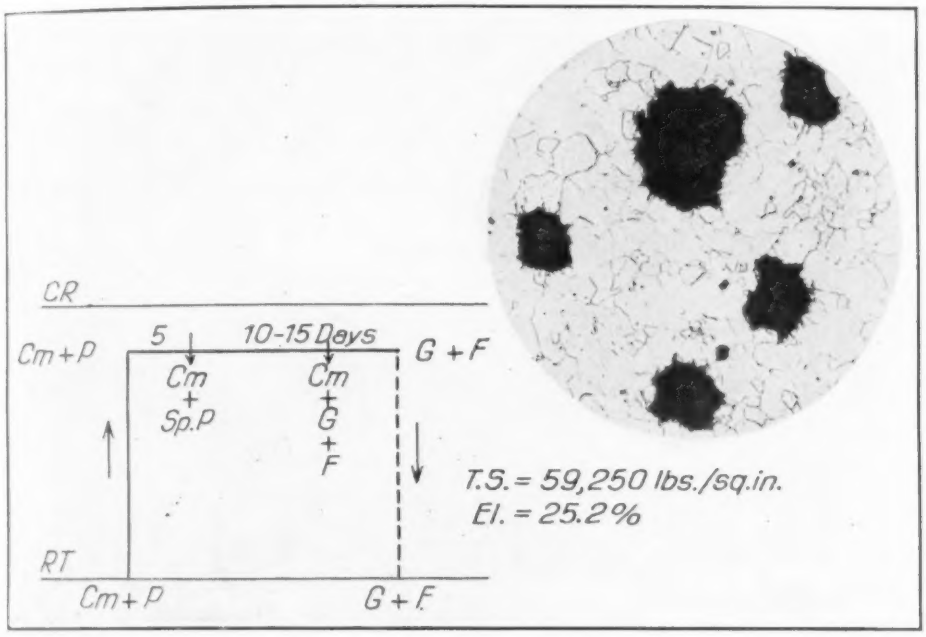


Fig. 5

elongation 19.1 per cent in two inches. These physical properties are superior to those possessed by ordinary malleable castings. There is some indication that the graphite G′ formed below the range is present in smaller particles. See Figs. 3 and 4 for comparison.

Complete Graphitizing Below the Critical Range

It should be possible to cause the complete graphitizing of cementite slightly below the critical range, although most likely the operation would require a very long time. This was attempted and accomplished as shown in Fig. 5. The temperature was 650 degrees Cent. (1200 degrees Fahr.), the time required between 10 and 15 days, the resulting tensile strength 59,250 pounds per square inch and the elongation 25.2 per cent. It will be noted that the ductility is substantially greater than that of malleable castings obtained in the usual way. This may be accounted for by the difference in the

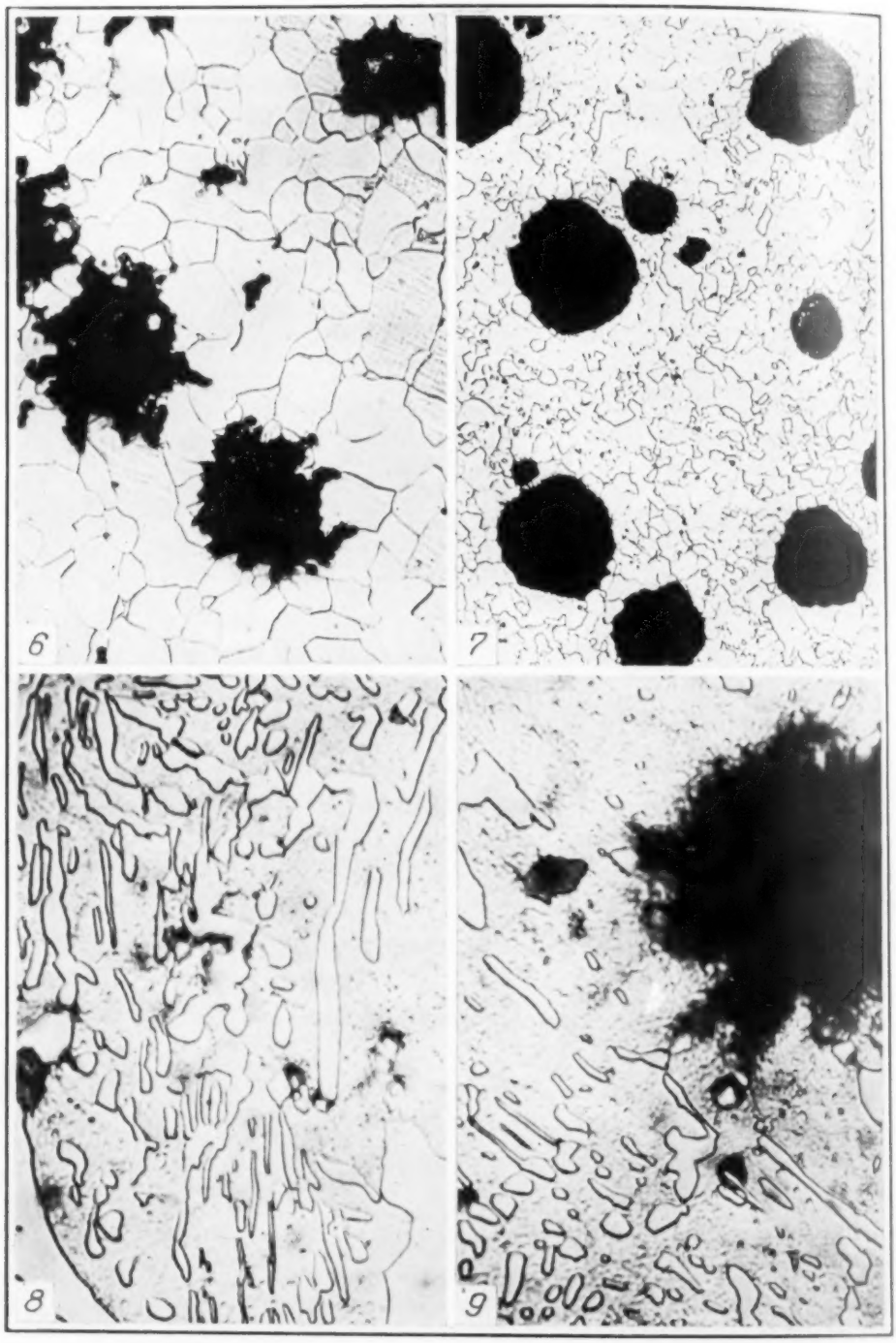


Fig. 6—Annealed 30 Hours at 870 Degrees Cent., 42 Hours at 890 Degrees Cent., 6 Hours Slow Cool to 770 Degrees Cent., 675 Degrees Cent. for 36 Hours, 18 Hours at 600 Degrees Cent. × 150.
Fig. 7—Annealed 14 Days at 650 Degrees Cent. × 150.
Fig. 8—Annealed 3 Days at 700 Degrees Cent. × 1500.
Fig. 9—Annealed 4 Days at 700 Degrees Cent. × 1500.

form and size of the graphite particles, as brought out in Figs. 6 and 7. The temper carbon in the ordinary malleable castings (Fig. 6) occurs in irregular particles, while when produced below the range (Fig. 7) it assumes a more rounded form. Again the ferrite grains after the usual annealing treatment are much larger than those obtained below the range. It is interesting to note how the graphite follows the boundaries of the ferrite in which it is embedded. The progress of graphitizing below the range may be followed through an examination of Figs. 8, 9, 10 and 11. It will be noted that after three days at 650 degrees Cent. (1200 degrees Fahr.) (Fig. 8) the pearlite has been completely spheroidized and that graphitizing has started. The sample represented by Fig. 11 was etched electrolytically.

Pearlitic Matrix

Malleable castings having pearlitic matrices and consequently greater strength may readily be obtained by cooling at a suitable rate after annealing above the critical range (Fig. 12). The casting should cool slowly enough through its critical range to permit the transformation of austenite into pearlite and quickly enough below the range to prevent the spheroidizing of the pearlite. The tensile strength was 83,500 pounds per square inch, and the elongation 2.24 per cent in 2 inches. It will be noted (Fig. 13) that each particle of temper carbon is surrounded by a substantial amount of ferrite. May this not be due to the fact that during slow cooling from the annealing temperature to the critical range some cementite forms which necessarily graphitizes, the pre-existing graphite particles serving as nuclei and the ferrite thus set free, remaining in situ surrounding them. Why this free ferrite should not be absorbed by the surrounding austenite is not clear, unless indeed the cooling is not sufficiently slow to permit this diffusion.

Sorbitic Matrix

The production of a sorbitic matrix should likewise be a simple operation. It suffices after annealing above the critical temperature to cool the casting through and below its critical range in such a way as to cause the austenite to be converted into sorbite—necessarily a faster cooling than required for the formation of a pearlitic matrix. This is illustrated in Fig. 14. The tensile strength is now 116,200 pounds per square inch and the elongation 3.1 per cent

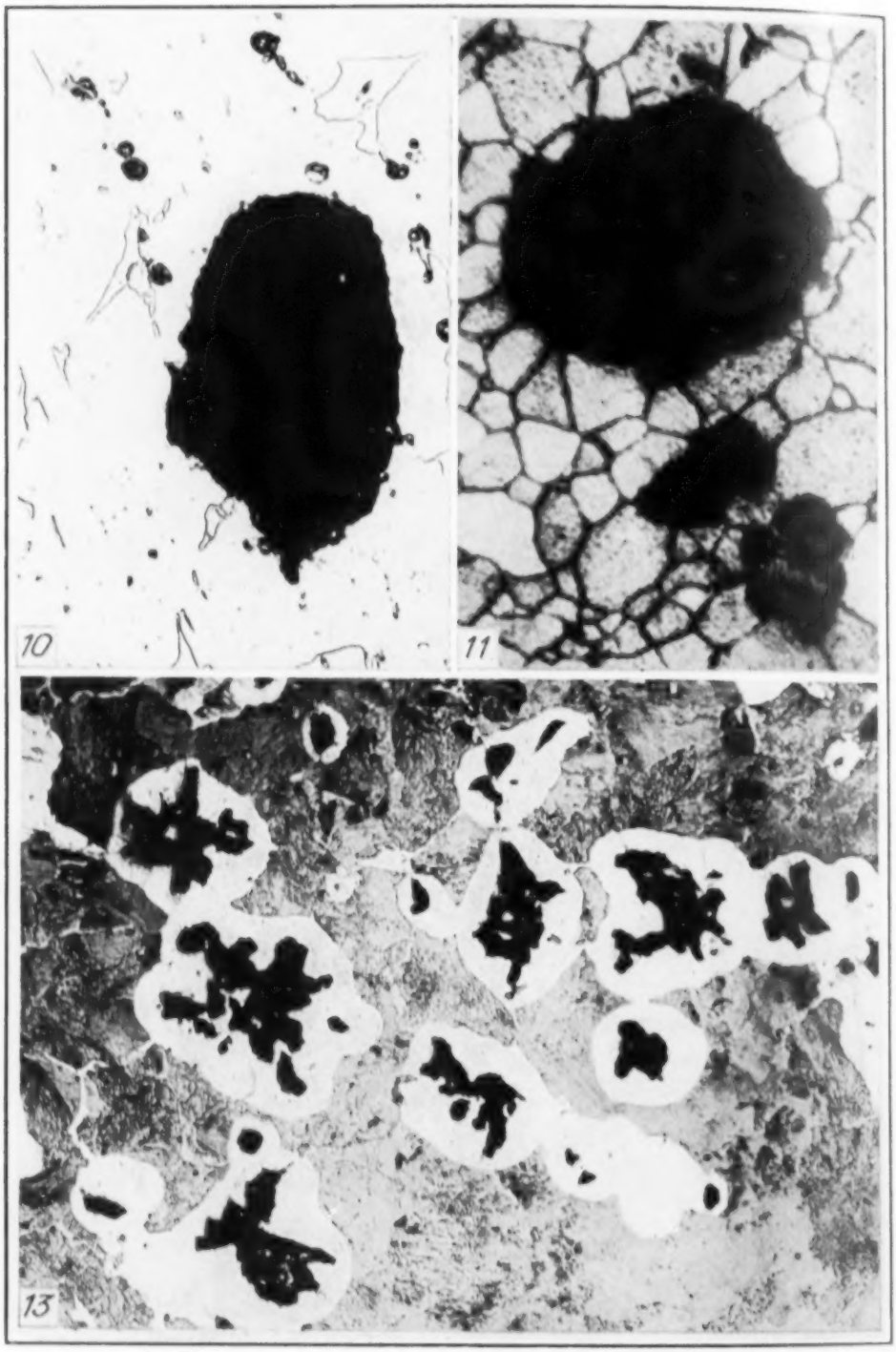


Fig. 10—Annealed 9 Days at 650 Degrees Cent. × 1500.
Fig. 11—Annealed 14 Days at 650 Degrees Cent. Electrolytically Etched. × 850.
Fig. 13—Annealed 24 Hours at 960 Degrees Cent. Slowly Cooled in the Furnace. × 150.

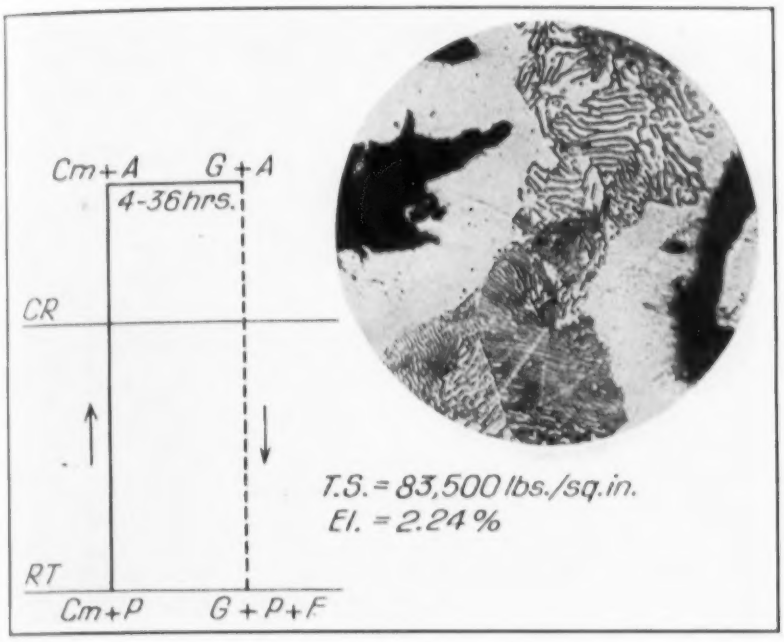


Fig. 12

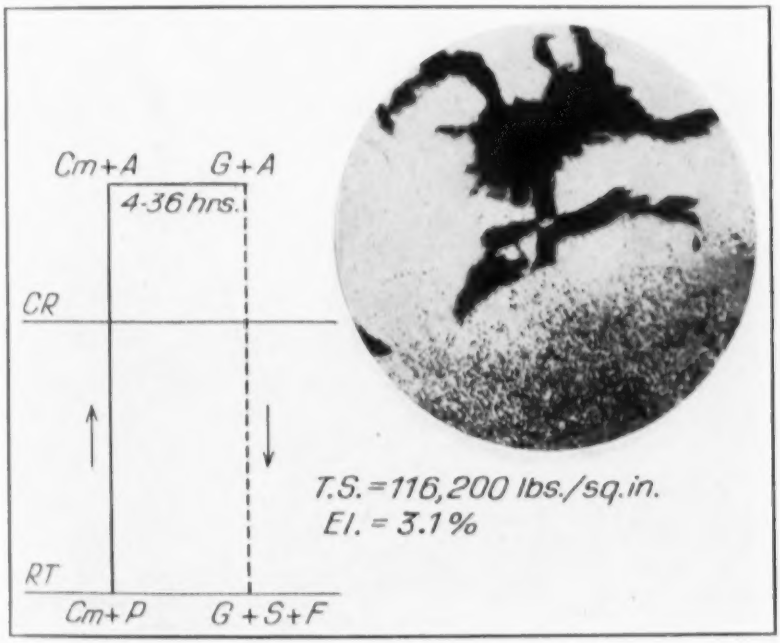


Fig. 14

in two inches. A sorbitic matrix may also be obtained by quenching the casting from the annealing temperature followed by reheating to some 650 degrees Cent. (1200 degrees Fahr.), Fig. 15. The physical properties are about the same.

Spheroidized Matrix

To produce a spheroidized matrix the castings after annealing above the critical range may be cooled slowly to some 650 degrees Cent. (1200 degrees Fahr.) where it should be maintained long enough to cause the pearlite to spheroidize, but not long enough to

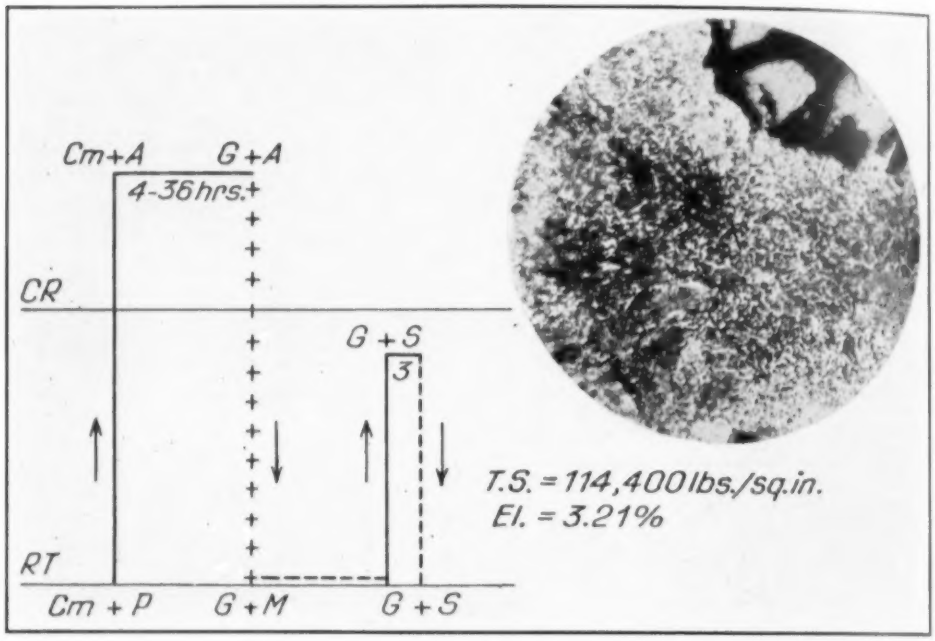


Fig. 15

permit its graphitizing. Or, preferably, it may be quenched from the annealing temperature, followed by reheating in the spheroidizing range, as shown in Figs. 16 and 17. The spheroidizing was obtained after 5 hours at 650 degrees Cent. (1200 degrees Fahr.). The tensile strength now reaches 123,200 pounds per square inch, while the elongation is 7.12 per cent. The total time for this treatment needs not exceed 40 hours. The casting is not so ductile as malleable iron can be made, but please note its strength.

Speed of Graphitizing at 650 Degrees Centigrade

The graphitizing of Fe_3C is accompanied by increased volume. This expansion may be taken as a measure of the graphitizing. A sample of white cast iron of suitable dimensions was placed in a Chevenard dilatometer, heated to 650 degrees Cent. (1200 degrees Fahr.) and maintained at that temperature for 26 days. The results

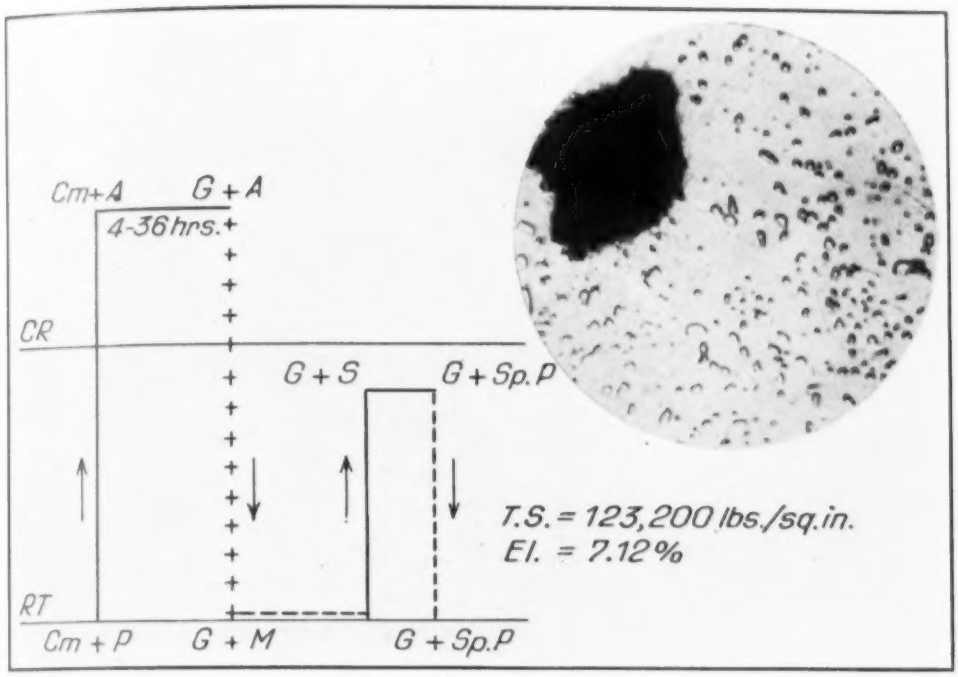


Fig. 16

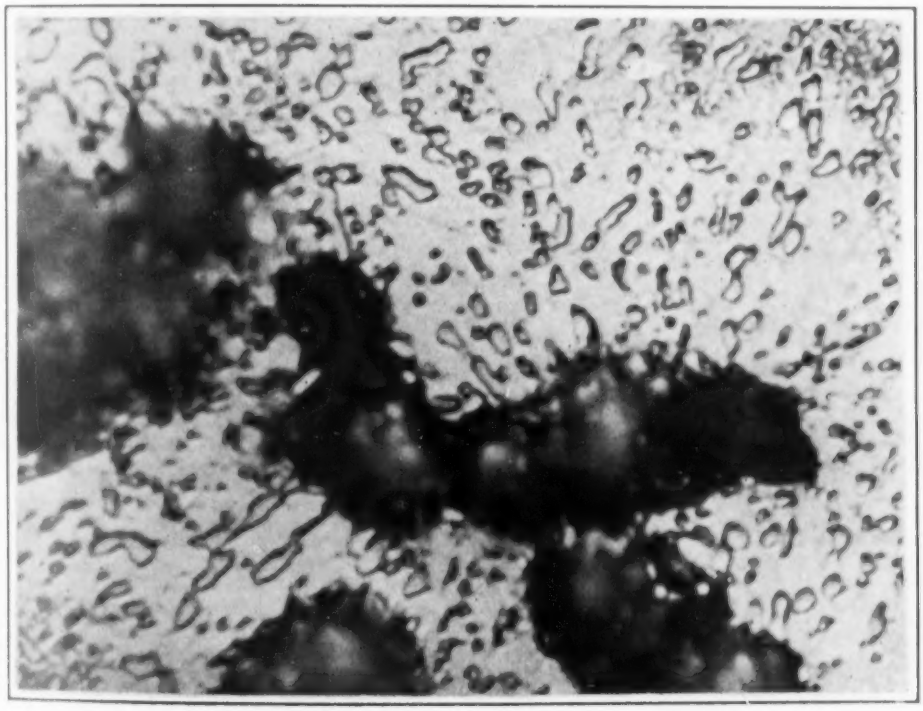


Fig. 17—Annealed 48 Hours at 900 Degrees Cent., Water-Quenched, Reheated at 650 Degrees Cent. for 5 Hours. × 2500.

are shown in Fig. 18. It will be observed that there was very little expansion during the first four days, hence little, if any, graphitizing.

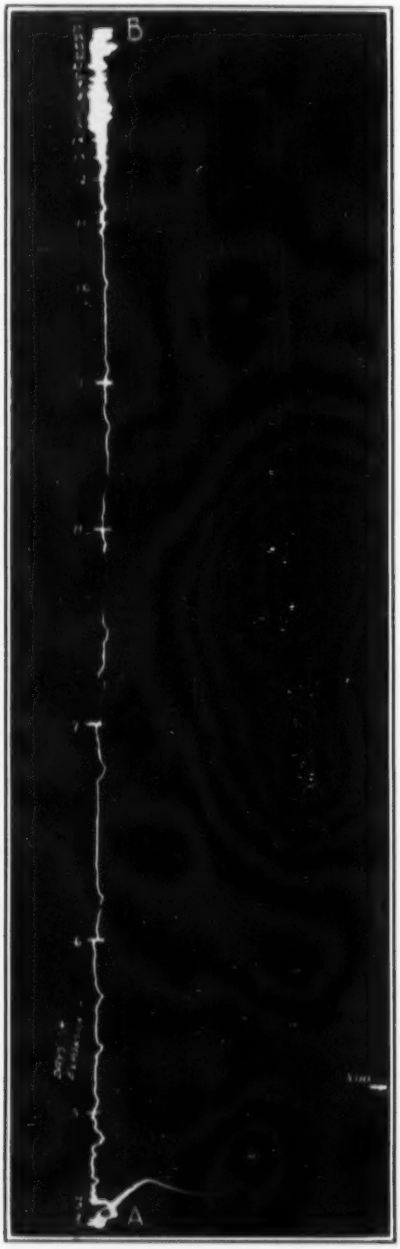


Fig. 18—Dilatometer Record for Heating Cycle of 26 Days at 650 Degrees Cent.

This period probably corresponds to the spheroidizing of the pearlite. A notable dilatation took place during the fifth day and then increased

in magnitude during the sixth and seventh day. A gradual decrease in the volume change is then indicated. It seems probable that after 12 days the graphitizing is practically complete. These results agree quite closely with microscopical examinations. The total expansion is represented by the distance AB, which measures about 10.66 inches. As the curve is magnified 300 times, it corresponds to an actual increase of length of about 0.035 inches and as the specimen was 1.95

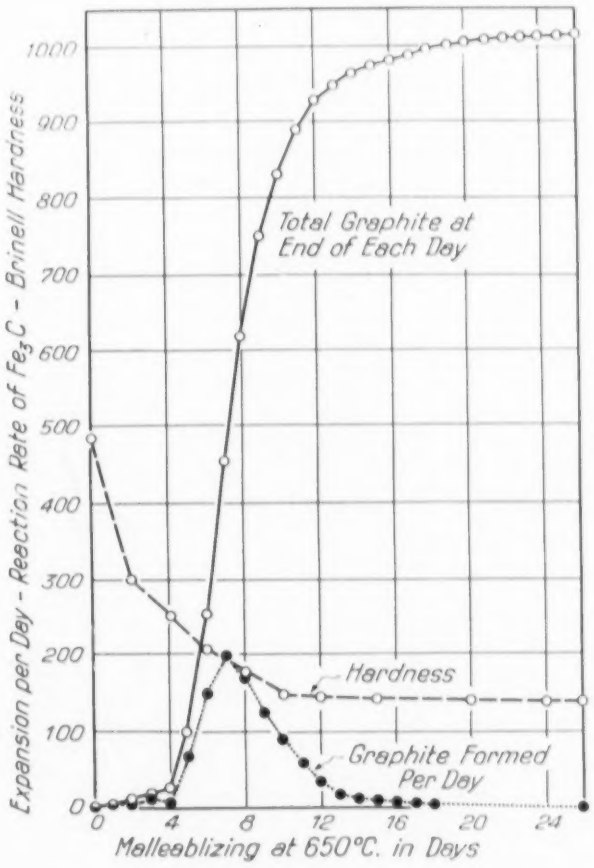


Fig. 19—Curves Showing Increase in Length and Brinell Hardness During Graphitizing.

inches long, it follows that its length increased 1.8 per cent during the malleableizing operation.

The curves of Fig. 19 indicate (1) the daily increases in length; (2) the total increase at the end of each day, and (3) the Brinell hardness at the end of each day. It will be noted that maximum softness was obtained at the end of the tenth day, although the specimen continued to expand. After fourteen days, however, the

expansion was very slight. It is assumed that the formation of graphite is directly proportional to the dilatation.

ACKNOWLEDGMENT

We desire to place on record our sincere appreciation to the following firms for having supplied us with the necessary white iron test bars—Columbis Malleable Castings Corporation, Providence, R. I.; Northern Malleable Iron Co., St. Paul, Minn.; Superior Steel and Malleable Castings Co., Benton Harbor, Mich.; Link-Belt Co., Indianapolis, Ind.; Albany Malleable Castings Corporation, Voorheesville, N. Y.; Badger Malleable and Manufacturing Co., South Milwaukee, Wis.; and Iowa Malleable Iron Co., Fairfield, Iowa.

DISCUSSION

Written Discussion: By H. A. Schwartz, manager of research, National Malleable and Steel Casting Co., Cleveland.

One is somewhat handicapped in the discussion of the paper by Sauveur and Anthony by its apparent abridgment from the original thesis. This has resulted no doubt in the omission of any reference to the literature and of much descriptive matter covering the basis of some of the conclusions. For example, those who have considerable contact with the difficulties of getting concordant results in the physical testing of cast materials are unable to judge whether the difference in physical properties reported in Figs. 3 and 5 are actually due to the difference in heat treatment or whether they represent accidental vagaries arising out of lack of perfection or lack of uniformity of test specimens. The experienced reader would value greatly further data and descriptive matter upon which he might base an individual opinion as to whether these differences of properties are well established in relation to the heat treatments or represent only the results which happened to be obtained.

It may be of interest to the authors to know that carrying out the entire annealing cycle below the critical point although at a temperature somewhat higher than 1200 degrees Fahr. was at one time quite common commercial practice. Thirty years ago many thousand tons of malleable were produced in such cycles. It may be said that the results were in general quite desirable so far as physical properties were concerned and to this extent the rule of thumb practice of that time confirms the present author's conclusions.

One finds it interesting to note that the authors regard it as necessary that the spheroidizing of pearlite must precede its graphitizing. That it frequently does so is a matter of common metallographic observation but enlargement on the view that it must do so, that is, that the lamina cannot graphitize directly would be greatly appreciated. This viewpoint is new to the commentator and an exposition of the reasoning upon which it is founded would no doubt be appreciated by many.

The form of graphitization curve presented by the authors is quite similar in its sigmoid form to that frequently recorded by others as for example, by Heyn, by Schutz and Stotz and by the present commentator. On the other hand, other observers, as, for example, Schneidewind, have considered that at least above the critical point the curve is logarithmic throughout. The commentator naturally welcomes the confirmation of his own past views. The sigmoid form of curve was observed both above and below the critical point. Above the critical point the flatness of the curve near the origin cannot be assigned to the spheroidization of pearlitic cementite. The author[1] developed a theory accounting reasonably well for the observed form of the curve by considering migratory rates; this investigation is here referred to only as pointing to the fact that a curve of the general form of Fig. 19 can be explained otherwise than by the method chosen by the present authors.

It may be well to point out for the benefit of other readers that there is considerable patent literature on the subject of such alloys as are here suggested as having higher strengths than ordinary malleable. These patents have been issued most extensively to Anson Hayes and his co-workers and in lesser number to Carl F. Lauenstein and to the writer and his associates.

The mechanism by which the well known bull's eye structure of Fig. 14 is formed is a field meriting further attention of interested metallographers.

Written Discussion: By N. A. Ziegler, Power Use Research Department, The West Penn Electric Co., Pittsburgh.

It is indeed gratifying to know that such an eminent scientist as Dr. Sauveur set to work on cast iron. I believe that I would be justified in saying that the mechanism of thermal phenomena occurring in plain carbon steels is quite clear to us, but we are by no means in a position to state the same about cast irons, for the simple reason that while the former for most practical purposes are binary alloys of iron and carbon, the latter cannot be considered as such.

When manganese, silicon and sulphur are present in the amounts as high as those given by Dr. Sauveur and Mr. Anthony, equilibrium conditions become very complicated: all points in the binary iron-carbon diagram become lines, and all lines, surfaces. The chemical compositions of the eutectic and eutectoid are also changed.

To the best of my knowledge, a constitutional diagram for iron-carbon alloys containing silicon, manganese and sulphur in the amounts usually present in ordinary cast irons, at present is unknown, and therefore the binary diagram has to be used as the nearest approximation. Piwowarsky did perhaps more than anybody else in solving this problem, but even his results are not conclusive.

The object of these remarks is merely to emphasize the lack of knowledge on the metallography of cast iron and the extreme value of any work aimed for illuminating this problem, particularly when it is carried out and presented as ably as that by Dr. Sauveur and Mr. Anthony.

Written Discussion: By W. J. Diederichs, metallurgist, The Autocar Co., Ardmore, Pa.

The paper by Dr. Sauveur and Mr. Anthony was of great interest to us, particularly the clear and concise method of presenting the results. It has per-

[1]H. A. Schwartz, "Graphitization at Constant Temperature," TRANSACTIONS, American Society for Steel Treating, Vol. 10, 1926, p. 53.

mitted the presentation of a great deal of information in a very short space; so short that the long hours and careful procedure required to obtain such data is apt to be forgotten.

During the past ten years, or more, there has been considerable discussion concerning the possibility of making short-time malleable, and that intermediate products of exceptional physical characteristics were theoretically possible was recognized. Those possibilities have been demonstrated by the authors in the clearest possible manner. As the authors state, there seems, however, to have been little desire until recently on the part of manufacturers to modify the matrices of the castings to obtain such properties.

We have been interested in the mechanism of graphitization over a period of many years and have accepted a theory to explain the mechanism. It is disconcerting to find that our understanding of the mechanism differs in a number of details from that of the authors. It does not coincide in the view expressed in the abstract of the paper that carbon in austenite must be thrown out of solution in the form of pearlite (or carbide?) before it can be graphitized, though under certain conditions this can and does happen. Neither does it coincide in the view, therefore, that the spheroidizing of pearlite is a necessary step preceding the graphitizing, though doubtless this precedes its graphitization if pearlite is once formed. Further, with reference to the authors' discussion of pearlitic matrix on page 415 and particularly the formation of a ferrite ring around each graphite nodule, the theory we have accepted indicates that the presence of the ring was not caused by prior cementite formation.

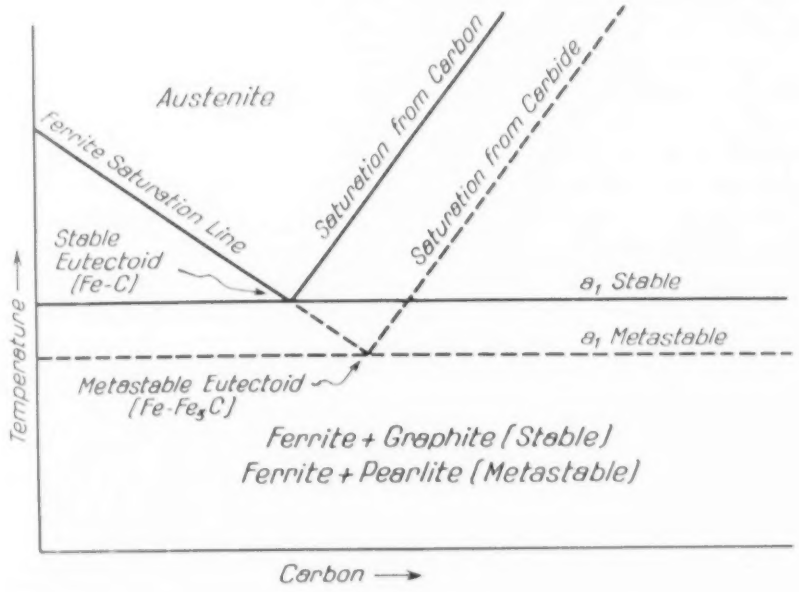
To repeat, it is disconcerting to find oneself at variance with Dr. Sauveur on a matter of theory. We beg leave, therefore, to present the theory we have accepted in order that we may derive the benefit of a discussion of it and, if it proves ill-founded, change our views accordingly.

It is quite generally accepted that throughout the temperature range employed in these experiments cementite is metastable; that the stable system is iron-carbon and the metastable system is iron-iron carbide; and this is particularly true when silicon approximates one per cent as in malleable castings. In such related systems, the critical temperatures of the stable system lie higher than those of the metastable system and the solubility of the metasable constituent is greater than that of the stable constituent. In the present instance, the saturation limit in gamma iron of carbon derived from carbide is greater than the saturation limit of carbon derived from carbon. The carbide saturation line lies to the right of the carbon saturation line in the constitutional diagram. Reference is made to Page 320 of The National Metals Handbook, 1933 edition. For purposes of discussion and enlarged section neighboring the eutectoid points of the figure there shown is given.

Upon heating a white iron above the critical range, the pearlite is converted to austenite in which is dispersed much massive carbide. Upon soaking at annealing temperature, this carbide goes in solution in the austenite until the austenite is saturated with carbon from carbide, attaining saturation first in the immediate neighborhood of the carbide particles. At this stage, it is supersaturated in respect to carbon from carbon and as a result carbon is deposited. Graphitization, as we see it, always starts at the surface of carbide particles

because supersaturation is first attained here. The solution of carbide and the deposition of graphite continues until all free carbide has disappeared and then deposition of graphite continues until the equilibrium concentration of carbon from carbon for the temperature is attained. At this stage we have austenite with carbon in solution to a greater concentration probably than either the stable or metastable eutectoid concentrations and in this austenite are dispersed nodules of graphite.

Upon slow cooling to the critical range, two things may happen: (1) if the cooling is sufficiently slow to substantially maintain equilibrium along the



carbon saturation line then the austenite is kept only slightly supersaturated and additional graphite will deposit on the nodules already present (seeding action), and no carbide will be formed. Eventually, the residual austenite will attain the concentration of the stable iron-carbon eutectoid. (2) If the cooling is somewhat too fast, the austenite may be undercooled to below the carbide saturation line and then carbide will be deposited; we believe always in the form of grain boundary cementite. The action is entirely analogous to the cooling of a hypereutectoid steel. If the cooling at this rate is continued to below the critical range, the austenite composition will follow down the carbide saturation line by deposition of grain boundary carbide until the metastable eutectoid is reached at which time the remaining austenite converts to pearlite. If the material is quenched from annealing temperatures, a martensite, troostite or sorbite hardening structure will be formed from the supersaturated austenite and grain boundary carbides may not be evident. This is again analogous to the results obtained on a hypereutectoid steel.

Upon cooling through the critical range, two things may again happen: (1) If the cooling from annealing temperature has been slow enough to permit austenite carbon concentration to maintain substantial equilibrium along the carbon saturation line, the austenite at a_1 stable temperature, or slightly be-

low, will transform to ferrite and graphite; iron-carbon eutectoid structure will result. This action is, however, very slow and at this stage the cooling must be very slow, or there must be a hold of temperature until sufficient heat has been abstracted. Sufficiently slow cooling here will complete the graphitization without prior formation of carbide or pearlite. (2) If the austenite has been brought to or near the concentration of the stable eutectoid, as outlined, it is hypoeutectoid with respect to the metastable eutectoid (pearlite point). It is now located also on the ferrite saturation line.

If the cooling from this last described condition is too rapid to permit the formation of iron-carbon eutectoid structure, then the remaining austenite acts exactly like the austenite of a hypoeutectoid steel; it deposits ferrite along the ferrite saturation line, enriching itself in carbon, until the concentration of the metastable eutectoid or pearlite point is reached, whereupon pearlite is deposited. It is to be noted, however, that an appreciable deposition of ferrite has occurred before the pearlite was formed. There is one distinction between the action of the deposited ferrite in this austenite and that of hypoeutectoid steel. In the steel, the areas of greatest instability appear to be at the grain boundaries and proeutectoid (metastable) ferrite deposits there. In this austenite, the greatest instability appears to be at the areas surrounding the graphite nodules and the proeutectoid (metastable) ferrite deposits around the nodule. This is possibly caused by the seeding action of the graphite nodules. The speed at which this ferrite ring forms is not surprising for ferrite deposition and austenite enrichment during the cooling of hypoeutectoid steels is quite rapid.

This is our understanding of the formation of the ferrite ring around graphite nodules (the bull's eye structure) and we believe it can be formed only in such cases where the cooling procedure down to the critical range has been such as to reduce the carbon concentration of the residual austenite to less than that of the metastable eutectoid. In the light of this theory we believe that the presence of the ferrite ring indicates that no carbide was present as such before the formation of pearlite at the metastable eutectoid.

It might be well to note further that any cooling which is started at too rapid a rate at any time before the concentration of carbon in the residual austenite has been reduced to below metastable eutectoid concentration will result in elimination of the ferrite ring and the structure will consist of nodules of graphite in a matrix of pearlite with, possibly, grain boundary cementite; or, the matrix will consist of one of several hardening structures, depending upon the particular cooling rate employed. This has been clearly demonstrated by the authors. As the authors have shown further, these matrix structures may be treated in various ways to develop various physical properties.

The observation of the authors that spheroidizing a sorbite matrix increases both strength and elongation is difficult to understand. An increase of elongation would, it seems to us, be expected, but it is difficult to understand the strength increase. We are led to ask whether the two samples of Figs. 15 and 16 are from the same heat of white iron and from the same batch of annealed and quenched samples. The photomicrograph of Fig. 15 shows a portion of a graphite nodule which appears more irregular in shape than the nodule in Fig. 16, as though the annealing temperature had been considerably higher. If the

analysis of the two samples differed somewhat and if the nodules of the sample of Fig. 15 are actually of poorer form or in greater amount than those of sample Fig. 16 it might explain the strength results noted.

In view of what has been said, the figures presented by the author would be of considerably more interest to us if the data as to particular annealing temperature and time, cooling rate to the critical range and cooling rate through the range had been given for each figure. We would be grateful to the authors, in case it is still available, if they would include such data in their closure.

Written Discussion: By Richard Schneidewind, Department of Engineering Research, University of Michigan, Ann Arbor, Michigan.

The authors have presented the results of a great deal of careful work which should be of considerable value to the malleable iron industry since heat treated malleable iron is becoming increasingly important. Some of the theoretical points, however, are not apparently in accord with the results of work done at the University of Michigan and it was felt that discussion might reconcile the differences. On page 412, the statement is made that "the cementite present in pearlite must necessarily spheroidize before it graphitizes. In Fig. 3 it is intended to indicate that it requires 15 hours to spheroidize the pearlite and nine hours more to graphitize it." In our work we found that in second-stage annealing the matrix was usually spheroidized, but not always. Occasional specimens remained lamellar almost to complete graphitization. In following the course of the graphitization reaction by chemical analyses rather than by dilatometric methods we never encountered the lag corresponding to the 15-hour spheroidizing time but found that graphitization started immediately and that the reaction followed the same laws as do any other supersaturated solutions.

On page 418, the authors note a similar lag in a specimen annealed entirely at 650 degrees Cent. (1200 degrees Fahr.) in 26 days. Their results are in accord with those of Kikuta who likewise employed the dilatometer.

The statement is made on page 421, "it is assumed that the formation of graphite is directly proportional to the dilatation." The formation of graphite means the formation of definite quantities or weights of graphite. If we assume a constant density for graphite, an increase in quantity will mean an equivalent increase in volume. But dilatation is the measurement of changes in linear dimensions and if the quantity or volume of graphite is increased K times, the linear dimensions will increase but $\sqrt[3]{K}$ times.

It would be interesting if the authors could parallel the graphite curve in Fig. 19 by one obtained through chemical analysis. There might be a lag between the formation of graphite and its dilatometric effect. The possibility is strengthened by appreciating that at the temperatures of annealing the metal is in the plastic rather than in the elastic state. In the plastic state there is a considerable lag between the time of application of load and the time at which total deformation results therefrom. It would be logical to believe, therefore, that the metal does not immediately expand due to the deposition of temper carbon. That the formation of graphite at constant temperature is accompanied by stress in the surrounding ferrite grains is indicated metallographically by the strain lines in these grains and indirectly by the fact that abnormal grain growth is frequently encountered if annealing is prolonged.

Authors' Closure—By H. L. Anthony

In reply to Dr. Schwartz's discussion pertaining to concordant results of physical data obtained from malleable produced at 650 degrees Cent., I have determined the physical properties of some 60 tensile test bars and data coincides very closely to that which is included in this paper. The differences in microstructure may be seen more clearly in Figs. 6 and 7 rather than in Figs. 3 and 5.

The only mention in literature to low temperature malleable that I have been able to find has been written by Dr. Schwartz as mentioned in his discussion, but the contents of his paper lacked physical properties, photomicrographs, and a description of the mechanism in the structure change during the graphitization of the white cast iron in the as-cast condition. The subject of prequenching from above the critical range before graphitizing at 650 degrees Cent. involves still another mechanism of graphitization which has not been mentioned in our paper because at the time of publication sufficient data had not been collected.

The discussion which Mr. Ziegler offers concerning the complexity of equilibrium conditions in iron-carbon alloys containing manganese, sulphur, and silicon has been appreciated.

Mr. W. J. Diederichs' discussion presents the mechanism of graphitization in the light of the double diagram, particularly in relation to the graphitization on cooling which includes the resulting structure formed in respect to the Ar_1 stable (iron-carbon eutectoid) and the Ar_1 metastable (iron-cementite) points.

In spite of the comments favoring the formation of the iron-carbon eutectoid, I am not sure that 0.97 per cent silicon content by slow cooling (10 degrees per hour) can possibly eject completely the carbon or Fe_3C contained in the austenite. That at temperatures slightly above Ar_1 stable, carbon or Fe_3C should have a certain solubility in the remaining austenite even at the mentioned rate of cooling. If a completely malleableized specimen is heated slightly above the A_1 stable for a short time there is a solubility for carbon, so why should not there be a solubility for carbon in austenite on slow cooling? If there is such a carbon solubility there is no reason why the cementite eutectoid is not formed at the transformation point.

I believe that possibly a certain amount of carbon precipitation can take place from the pearlite below the Ar_1, but the more favorable state being in the spheroidized condition.

Of course it is known that silicon decreases the solubility of carbon in austenite and shifts the eutectoid point to the left, but whether the presence of less than 1.00 per cent of silicon through suitable cooling rates can form a hypoeutectoid matrix from above the Ar_1 point is a matter of discussion. Thus, the presence of ferrite rings adjacent to the graphite has been difficult to explain.

If quenching a series of specimens from points just above the Ar_1 point, martensite results, we cannot say that ferrite has been liberated at the graphite interface during this slow or rather rapid cooling. On the other hand the ferrite rim may be formed at the point of transformation. This can be explained more clearly if the annealing specimen is cooled to the Ar_1 and then quenched in a molten bath of aluminum at 675 degrees Cent. followed by water quench-

ing; the pearlite extends in most cases to the graphite interface and the ferrite rim does not appear.

In studying graphitization by means of the dilatometer, a decided expansion occurs on cooling through the Ar_1, indicating that a decomposition of this nature may possibly take place.

I am unable to state definitely the difference in the annealing temperatures of the specimen used in Figs. 15 and 16, but were in the range of 900-960 degrees Cent. The magnifications of these two photographs are different, Fig. 15 being at 800 and Fig. 16 at 2000 diameters. The difference in the character of the graphite is due to the fact that at high magnifications a complete nodule of graphite could not be photographed when only a limited printing area was available, so only a portion of the nodule was included in Fig. 16. The rate of cooling for complete malleabilization was set at approximately 18 degrees per hour from 870 to 675 degrees Cent.

In regard to the Schneidewind discussion especially concerning the statement, "The cementite present in pearlite must necessarily spheroidize before it graphitizes." I might again mention that at the Ar_1 transformation a marked expansion is recorded on the dilatometric curves indicating the formation of graphite immediately from the cementite eutectoid, but after the transformation, the dilation curve approaches the horizontal for some 12-15 hours, after which, the expansion increases. However, it is possible that some of the cementite of the pearlite lamellae decomposes, but surely, the spheroidized form is the most favorable to complete the second stage reaction. On the other hand, it is much more definite that when graphitization takes place at 650 degrees Cent., the cementite of the pearlite must be in the spheroidized condition before graphitization can take place.

The curve in Fig. 18 was only employed to study the reaction rate at 650 degrees Cent.; by that I mean it was used to determine whether the decomposition of the cementite was the same on each succeeding day. From this it was found that the first three days the reaction was slow in starting which was simultaneously verified by microscopic examination which showed that during this period the pearlite was in the process of spheroidizing. If the specimen is prequenched so that we start with a martensitic matrix instead of a pearlitic matrix, the reaction is much more rapid, reducing the low temperature cycle at 650 degrees Cent., to approximately one half of the time required to graphitize the pearlitic matrix.

Reply to the Discussion by Dr. Albert Sauveur

In regard to Dr. Schwartz's remarks, it was quite intentional that the authors abstain from referring at any length to the literature dealing with the subject they are investigating, as they preferred to have their conclusions stand unassisted upon the results obtained by them.

It is noted with interest that at one time it was quite a common commercial practice to carry the entire annealing cycle below the critical point. The authors confess their ignorance of that fact.

In regard to the spheroidizing of pearlite necessarily preceding its graphitizing, that seems to us quite obvious. In our opinion, the graphitizing of pearlite calls for a long annealing slightly below the thermal critical range,

which is also a spheroidizing treatment. It is not seen how the spheroidizing stage could be skipped.

We believe that our explanation of our curve, shown in Fig. 19, is simpler than the explanation to which Dr. Schwartz refers, while it is logically as acceptable. Hence it is preferable.

Our attention is called to the patent literature on the subject under investigation. Perhaps it would be wise to make a search of the patent office file before undertaking any research. The task, however, might well prove formidable.

We fully agree with the views of Mr. Ziegler regarding our lack of knowledge of the equilibrium conditions of so complex an alloy as is represented by an association of iron, carbon, silicon and sulphur. He will note that we did not venture into that field.

The very able and clear discussion of Mr. Diederichs is much appreciated. His entire argument must stand or fall on the possibility of graphite forming directly from austenite without passing through the intermediate stage of Fe_3C. We admit that graphite is the stable form, but it does not follow that carbon formed directly. We admit the reasonableness of the double diagrams, but that does not prove its accuracy. We believe that evidences are still lacking pointing to the separation of carbon atoms from the space lattice of gamma iron in order to form particles of graphite.

A GRAPHICAL METHOD OF CALCULATING HEAT LOSS THROUGH FURNACE WALLS

By Clarence E. Weinland

Abstract

When a wide range of temperatures is considered, difficulty is encountered in accurately representing the thermal conductivity of any material, the temperature-conductivity function of which is not a straight line relation. This difficulty can be obviated by use of the "true" conductivity function as defined herein.

The use of the "true" conductivity function confers the additional advantage that it makes possible a graphical method for the calculation of heat transmission through compound structures, under conditions of equilibrium heat flow. Though this is still a trial and error method, it is much more convenient and direct than any method heretofore proposed for handling this type of problem. The method is illustrated by the solution of a typical problem.

Introduction

IN order to calculate the heat transfer through any structure under steady temperature conditions, the first requisite is a knowledge of the thermal conductivity, or heat transmitting ability of unit volume of each of the materials of which the structure is composed. When wide ranges of temperature are to be considered, as they inevitably must be in the design of metallurgical furnaces, all of the materials used in furnace construction are found to change markedly in thermal conductivity, when each is taken through its normal range of operating temperatures.

The question of how best to represent the thermal conductivities of the materials of furnace construction, and how to use them in calculation, has appeared from time to time in engineering literature. McMillan[1] discussed the subject and concluded that the aver-

[1] L. B. McMillan, "Heat Transfer Through Insulation in the Moderate and High Temperature Fields," *Transactions*, American Society of Mechanical Engineers, Vol. 48, 1926, p. 1269.

A paper presented before the Sixteenth Annual Convention of the Society held in New York City October 1 to 5, 1934. The author, Clarence E. Weinland, is physicist for Johns-Manville, Manville, New Jersey. Manuscript received July 24, 1934.

age conductivity of a material between two temperatures should be expressed as a function of the mean temperature. This method of expressing conductivity is satisfactory when the conductivity-temperature function for a material is a straight line relation, or has but little curvature; otherwise its accuracy is open to question. This difficulty is intensified when large temperature differences are encountered. To go into a mathematical analysis here of the magnitude of the errors introduced by the use of the mean-temperature average-conductivity function would be beside the point and would require too much space.

In addition to the difficulties encountered in accurately stating the conductivity of any material, there has generally been more or less difficulty involved in the use of thermal conductivity data in the calculation of heat losses through walls consisting of several layers of different kinds of material, through which the heat must flow successively. Such a problem usually is a matter of predicting the behavior of a design under consideration, and the available data nearly always included the interior furnace temperature, which will be assumed equal to the temperature of the inner face of the furnace wall throughout this discussion, the outer ambient air temperature, and the thickness and mean-temperature conductivity relation for each of the materials of which the wall is to be constructed. Without data at hand regarding the temperatures at the interfaces between materials in the design, it is not possible to obtain directly the mean temperature of any of the component materials, and, consequently, the correct value of conductivity to be assigned to each material is unknown, even though the conductivity-temperature data are at hand. The most satisfactory method of attacking the problem has probably been the method of successive approximations, starting with rough assumptions as to interface temperatures and conductivities, and refining the assumed values for each new calculation until a final set of data has been obtained which is consistent with the conditions of the problem. This method is tedious, and requires familiarity for satisfactory use. Trinks has given an example of this type of calculation.[2]

The Basis for the Present Method

In order to make a new approach to the problem, it is necessary to go back to the fundamental mathematical equation for steady state

[2]W. Trinks, "Industrial Furnaces," Vol. I, 2nd Edition 1926, page 67, John Wiley & Sons, Inc., N. Y. City.

heat flow, set up by Fourier in 1822. This may be given the form

$$dq = k\,dt \qquad (1)$$

where (dq) is the amount of heat which will flow through a unit cube of any material when the temperature difference between its hot and cold faces is (dt) and its thermal conductivity is (k). It is evident that if the temperature difference between opposite faces of the cube is taken small enough, the variation of thermal conductivity (k) with temperature within the cube becomes negligible, even for purposes of rigorous analysis and the conductivity of the material (k) at the temperature of the cube (t) becomes a definite unambiguous quantity. If now the temperature (t) of the cube is allowed to vary, it is possible to state the thermal conductivity as a function of each different temperature value throughout the useful working range of the material. This concept of thermal conductivity will be referred to as the "true" thermal conductivity (k) hereafter, to distinguish it from the average conductivity taken over a wide temperature interval. Recent developments[3, 4] in the practice of thermal conductivity measurement are making it possible for us to obtain relatively good approximations to the true conductivity functions for the various materials of furnace construction, by testing them under small temperature differences or by suitable mathematical treatment of test results obtained.

A method for graphically calculating the heat loss through a furnace wall consisting of a refractory and an insulating material has been given by Trinks.[5] This method, however, has the distinct disadvantage that a set of curves such as those used are not applicable to more than a limited range of conditions; the method of constructing the curves involves the use of mean thermal conductivities with all their inherent uncertainties; and the practice of the method is a more or less tedious process of adjusting the relative positions of three curves, one of which must be drawn on a transparent sheet.

The use of the true conductivity function in the calculation of

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[3] G. B. Wilkes, "The Thermal Conductivity of Refractories," *Journal*, American Ceramic Society, Vol. 13, 1934, p. 173.

[4] C. E. Weinland, "A Method of Measuring Thermal Conductivity at Furnace Temperatures," *Journal*, American Ceramic Society, Vol. 17, 1934, p. 194.

[5] W. Trinks, "Industrial Furnaces," Vol. I, 2nd Edition, 1926, p. 69, John Wiley & Sons, Inc., N. Y. City.

heat losses was indicated in a limited way by Langmuir[6] and by Randolph.[7] They applied their method to the calculation of heat losses in one case through a gas film around a heated filament, and in the other case through a single layer of thermal insulation surrounding an electric stove heating element. Apparently they made no attempt to apply their calculating method to heat flow through two or more materials in series.

The graphical method of calculation to be described here derives its fundamentals from the method of Langmuir and Randolph, and so is based upon the true conductivity functions for the materials of construction. Equation (1) refers to the heat flow through a unit cube of any material with an exceedingly small temperature difference between its hotter and cooler faces. For any section, under steady temperature conditions, the heat flow will be given by

$$dq = \frac{A}{L} k\, dt \qquad (2)$$

where (dq), (k) and (dt) are rate of heat flow, true thermal conductivity and the differential of temperature as before, and (A) and (L) are the area of the section and length of path of heat flow, respectively. Throughout this discussion it will be assumed that a section of furnace wall one square foot in area is being considered, and that all isothermal surfaces are parallel planes, so that the length of path of heat flow (L) is equal to the thickness of the material in inches. In the special case of radial heat flow in a cylindrical section of material, the factor (L) may be taken as the logarithmic mean thickness

$$L = r_s \log_e \frac{(r_2)}{(r_1)} \qquad (3)$$

where (r_2) is the outer and (r_1) the inner radius of the section, and (r_s) is the radius of the outermost surface of the construction, which then enables the determination of the heat flow through such a section, per unit area of outermost surface. This quantity may

[6] I. Langmuir, "Convection and Conduction of Heat in Gases," *Physical Review*, 1st series, Vol. 34, 1912, p. 401.

[7] C. P. Randolph, "The Conduction of Heat," *General Electric Review*, Vol. 16, 1913, p. 120.

be used in place of (L) in any calculation involving the charts to be described subsequently.

In order to determine the heat flow for a section of material of unit area, between any given temperatures on the hot and cold surfaces, equation (2) must be integrated

$$q = \frac{A}{L} \int_{t_c}^{t_h} k\,dt \qquad (4)$$

which may be separated into two simple integrals

$$q = \frac{A}{L} \left[\int_{o}^{t_h} k\,dt - \int_{o}^{t_c} k\,dt \right] \qquad (5)$$

If there is available a simple means of determining values of the integral

$$\frac{qL}{A} = \int_{o}^{t} k\,dt \qquad (6)$$

for any value of temperature (t), it then becomes a simple matter to determine heat flow per unit area, $\left(\frac{q}{A}\right)$, for any thickness of material (L), by looking up values of the integral corresponding to hot and to cold face temperatures of the material, subtracting one from the other and dividing the result by the thickness (L).

Langmuir[6] and Randolph[7] drew curves connecting values of the integral equation (6) and temperature (t), but it has been found much more convenient for the purposes of this work to plot the relation in the form of a function scale.[8] An example of such a scale may be seen in Fig. (1b), which represents the function

$$\frac{qL}{A} = \int_{o}^{t} k\,dt \qquad (7)$$

for a commercial insulating material. The true conductivity (k), versus temperature function of this material as measured in thermal conductivity testing equipment working with temperature differences of 100 to 150 degrees Fahr. across the test specimen is shown in

[8] L. I. Hewes and H. L. Seward, "The Design of Diagrams for Engineering Formulas," 1st Edition, 1923, p. 1, McGraw-Hill Book Co., N. Y. City.

Fig. (1a). The integration of this curve may be conveniently carried out by assuming it to be divided into sections covering 100 degrees Fahr. each, and by assuming further that the average conductivity of each such section is adequately represented by the true

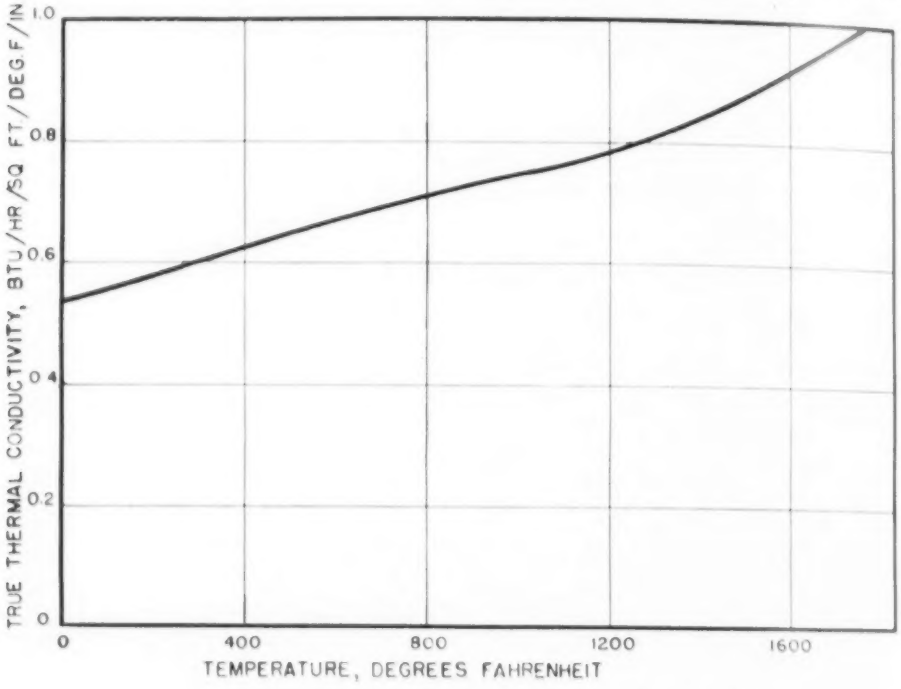


FIGURE 1A

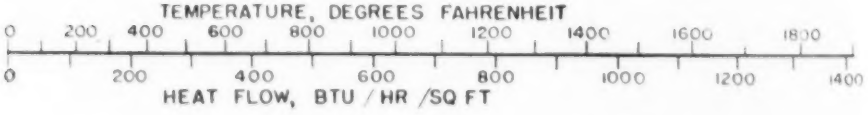


FIGURE 1B

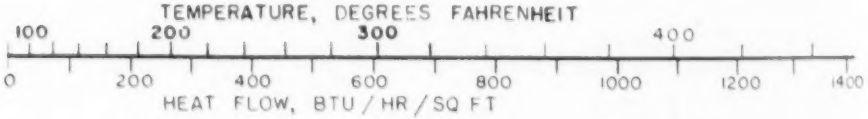


FIGURE 1C

Fig. 1A—True Thermal Conductivity-Temperature Relation, for a Commercial Insulating Material.
Fig. 1B—Function Scale for the Insulating Material, the Conductivity of which is shown in Fig. 1A.
Fig. 1C—Function Scale for the Surface Coefficient.

conductivity taken at the 50-degree point in the section. Each conductivity value is then multiplied by the temperature interval (dt) which it represents, that is by 100 degrees Fahr., giving a series of values of (k dt). Then starting with that interval the lower tem-

perature of which is 0 degrees, a series of sums of (k dt) values is made, each sum including one more interval than the last, and each one including all values of (k dt) between 0 degrees and the temperature value, (t^0), which forms the upper limit of the last interval included in that sum. This series of sums, with the related values of (t) constitutes the function represented in equation (7). To plot the function scale, it is then necessary only to draw a line, graduate it on one side with a uniform scale of values of $\left(\frac{ql}{A}\right)$, and on the other side to locate the values of the (k dt) sums just secured, labelling each such point with the temperature value of the upper limit of its range. Suitable temperature sub-divisions may then be inserted and the function scale is complete. To use such a function scale for the calculation of the heat flow through unit area and thickness of the material it represents, it is only necessary to locate the temperatures of hot and cold faces on the temperature side of the scale, read off the corresponding $\left(\frac{ql}{A}\right)$ or heat flow values, subtract the less from the greater and the difference will be equal to the desired heat flow value. For thicknesses of material greater or less than unity, multiply the heat flow value by the reciprocal of the thickness.

The Use of Function Scales in Combination

The use of function scales in combination will be illustrated first by calculation of a simple problem involving an insulating material, the scale for which is shown in Fig. 1B and the scale for the surface loss coefficient, Fig. 1C. The latter has been made up from average data for the loss from a large vertical plane surface of high emissivity, the ambient air temperature being 80 degrees Fahr., and the heat loss values being the total loss by radiation and conduction-convection.

For simplicity it will be assumed that the thickness of the insulation is one inch, that the inner temperature of the insulation is 1000 degrees Fahr. and that the air temperature is 80 degrees Fahr. Since the amount of heat which passes through one square foot of the insulation in unit time must be exactly equal to the

amount lost from one square foot of the outer surface, the correct solution of the problem will consist of a length of the scale for the insulating material extending from the 1000-degree Fahr. graduation down to some lower temperature point, (x^o), which will be the temperature of the outer surface of the insulation, and the same length of the surface coefficient scale extending from the room temperature, 80 degrees Fahr., up to the same surface temperature, (x^o). This length can be quickly obtained by setting a pair of dividers to some estimated length and applying it first to one scale and then to the other, and if the above condition is not fulfilled, resetting it and applying it again until the proper setting is found. In this case the surface temperature will be found to be 270 degrees Fahr., the distance on the insulation scale between 1000 degrees Fahr. and 270 degrees Fahr. being the same as that on the surface coefficient scale between 270 degrees Fahr. and 80 degrees Fahr., and being the same as the distance on either scale between the 0 and the 490 B.t.u. per hour graduations. The correct solution of the problem has then been obtained, and is a heat loss of 490 B.t.u. per hour per square foot, and a surface temperature of 270 degrees Fahr.

This is evidently a trial and error method, but instead of requiring a more or less lengthy calculation to verify each estimate a quickly made divider setting suffices.

Two scales are shown in Fig. 1, but in order to make up a chart which will be applicable to a wide range of furnace wall calculations it becomes necessary to add scales for other materials frequently used in furnace construction. Such a chart is shown in Fig. 2. In making up this chart, however, several simplifying features become desirable.

In the first place, it is not necessary to graduate each scale with the uniform heat loss divisions, for they can be included once for all on a separate scale as seen in the lowest line of Fig. 2. When the correct divider setting has been obtained the dividers may be applied to this heat loss scale and the final value of heat loss read off.

Secondly, when a number of scales for materials of widely different conductivity are brought together, they are found to have very different lengths. To overcome this difficulty each scale is expanded or contracted by a suitable "scale factor," so that it approximately fills the available space. In order to still make use of the single separate heat loss scale it then becomes necessary to divide B.t.u./Hr./Sq.Ft. readings from any one scale by the scale factor

Fig. 2—Chart for the Graphical Calculation of Heat Loss Through Furnace Walls.
Copyright 1934 by Johns-Manville Corporation

in order to find the true heat flow, or conversely if a known heat loss value is to be applied to the scale to find temperatures, it must be multiplied by the scale factor, and the divider setting must be made for the product of heat loss value and scale factor, on the heat loss scale before applying the dividers to the function scale for the material.

The third simplifying feature permits one to use any one of the function scales to represent any thickness of material. For if a given distance on any of the scales represents the amount of heat which would flow if the faces of unit thickness of the material were held at two given temperatures, then for greater thickness the distance on the heat flow scale must be divided by the product of the thickness of the material in inches and the scale factor; and, conversely, if a given amount of heat is known to be flowing through a section of thickness greater than unity, the value of heat flow may be multiplied by the product of the thickness in inches and the scale factor, and the length along the heat loss scale at the bottom of the chart corresponding to this product may be applied with the dividers to the function scale for the material in order to find the hot and cold face temperatures which would prevail.

The description of these three simplifications may make them seem to be only additional involvements, but all three contribute decidedly to the simplification of the form of the chart itself, while the second and third of them contribute markedly to the accuracy of the method, by making it possible to work at all times on scales of maximum length and openness.

The use of the complete chart, Fig. 2, may now be illustrated by solving a problem involving heat flow through a composite furnace wall. Assume the wall to consist of 13½ inches of fireclay brick, 4½ inches of C-22 Brick, and 3 inches of Superex insulation. Assume the temperature of the inner face of the firebrick to be 2400 degrees Fahr., and that of the outside air to be 80 degrees Fahr. For each material in the wall, thickness value (L) and scale factor (F_s) will be constant throughout the problem, so they may be combined into one "working factor" for each material as follows:

Material	Thickness (L)	Scale Factor (F_s)	Working Factor
Firebrick	13½	0.060	0.81
C-22 Brick	4½	0.250	1.125
Superex	3	0.700	2.10
Surface Coefficient	1 (arbitrary)	0.500	0.50

As an aid in forming a first estimate of the heat flow, the median thermal resistivity (1/k) for each material has been included on the chart. These values of resistivity multiplied by the thickness of each component will give the approximate resistance of each, and the sum will be the approximate total resistance, as follows:

Material	Resistivity 1/k	Thickness L, Inches	Resistance Deg. F./B.t.u./ Hr./Sq. Ft.
Firebrick	0.095	13½	1.27
C-22 Brick	0.500	4½	2.25
Superex	1.300	3	3.90
Surface Coefficient	0.400	1	0.40
		Total	7.82

The total resistance divided into the total temperature drop will give a first estimate of heat flow:

$$\frac{2400 - 80}{7.82} = \frac{2320}{7.82} = 297 \text{ B.t.u./Hr./Sq.Ft.}$$

In order to test this first estimate it is multiplied by each of the working factors in turn, and the dividers, set to each product, are applied to the function scales in turn, giving the temperature values which would obtain if that amount of heat were actually passing through. For firebrick $297 \times 0.81 = 241$ and the divider so set indicates a temperature drop from 2400 to 2020 degrees Fahr. For the C-22 brick $297 \times 1.125 = 334$ and the dividers indicate a temperature drop from 2020 to 1400 degrees Fahr. For the Superex $297 \times 2.10 = 624$ and the dividers indicate a drop from 1400 to 110 degrees Fahr. It is then evident that the estimate of heat flow was too high, for when the divider setting for the surface coefficient is applied, $297 \times 0.50 = 149$, with one divider point at 110 degrees Fahr. on the surface coefficient scale, the other point falls well below the given ambient temperature.

In order to assist in refining the estimates of heat loss, an error scale has been plotted along the lower side of the surface coefficient scale, with its zero coincident with the 80 degrees Fahr. ambient temperature graduation. It has been found empirically that a fair approximation to the necessary correction can be made by taking the temperature difference between the position of the final divider point and the given ambient temperature, halving it, and dividing

by the originally estimated overall resistance. The error scale just referred to has been laid out with uniform divisions each equal to half the temperature divisions of the surface coefficient scale in the neighborhood of 80 degrees Fahr., so the correction to a trial value of heat flow will be equal to the reading of the divider point on the error scale, divided by the approximate thermal resistance.

In the example under consideration, the final divider point indicated a position to the left of the 80 degrees Fahr. graduation on the surface coefficient scale, coinciding with the 60-degree Fahr. graduation on the error scale. The correction to the first estimate of heat flow is then $\frac{60}{7.82} = 7.7$, or about -8 B.t.u. per hr. per sq. ft. The new estimate of heat flow is, therefore, 297-8 or 289 B.t.u. per hr. per sq. ft.

This second estimate may be tried by repeating the foregoing process of application of the dividers, with results as indicated in the following table:

Material	Working Factor	Working Factor × Heat Flow	Hot Surface Temp., Degrees Fahr	Cold Surface Temp., Degrees Fahr.
Firebrick	0.81	234	2400	2030
C-22 Brick	1.125	326	2030	1424
Superex	2.10	607	1424	185
Surface Coefficient	0.50	144	185	About 55

The reading on the error scale is now only 15 degrees Fahr., and if divided by the resistance, 7.82, gives a correction of -1.9 B.t.u. per hour per sq. ft. Since this correction is less than one per cent of the estimated heat loss, reasonable accuracy will not be sacrificed if the value so corrected is taken as the final result of the calculation. It is desirable, however, that the cold surface temperature of the Superex be corrected, by applying the dividers again to the surface coefficient scale, with one point at 80 degrees Fahr. and spread to $287 \times 0.5 = 143$, giving a temperature value of 208 degrees Fahr. for the outer surface of the wall. Corrections to the other temperature values would in this case be negligible. The final solution of the problem, then, is:

Heat Loss, 287 B.t.u. per hour per sq. ft.	
Temperature, Hot Surface of Firebrick	2400 degrees Fahr.
Temperature between Firebrick and C-22	2030 degrees Fahr.
Temperature between C-22 and Superex	1424 degrees Fahr.
Temperature, Cold Surface of Superex	208 degrees Fahr.
Temperature of Surroundings	80 degrees Fahr.

This solution of the problem may be roughly checked by taking the thermal conductivity of each material from the conventional thermal conductivity-temperature curves, for each mean temperature as determined by the solution, and computing the heat flow through each component of the wall separately, excepting that the heat flow through the surface air layer may be taken directly from the surface loss curve.

Material	Mean Temp., ° F.	Thermal Conductivity, B.t.u./hr./sq. ft./ °F./in.	Temp. Diff., degrees Fahr.	Thickness, inches	Heat Flow, B.t.u./hr./sq. ft.
Firebrick	2215	10.50	370	13½	288
C-22	1727	2.17	596	4½	288
Superex	805	0.705	1216	3	286
Surface Coefficient	...	....	128	...	288

The chart shown in Fig. 2 has been made up to represent the heat transmission characteristics of seven kinds of material used in furnace construction. Natural Sil-O-Cel brick has a laminar structure and, consequently, has different characteristics depending upon whether the direction of heat flow is perpendicular to the laminations or parallel with the laminations, and so is respected by two function scales.

The scales for fireclay and silica brick have been made up from the average thermal conductivity data for each material as given by Trinks.[2] Both materials vary widely in thermal properties, depending upon composition and manufacture. Trinks' averages are probably the best available at present.

C-22 Brick is a calcined diatomaceous earth product, while Superex is a block material compounded of diatomaceous earth and asbestos fiber. C-3 concrete is mixed and formed on the job, consisting of a calcined diatomaceous earth aggregate with Portland cement binder.

The scale for the surface coefficient has been made up from average data for the heat loss from a large vertical plane surface with high emissivity, and zero wind velocity. Should it be desirable to take account of wind velocity, it can be done by arbitrarily assigning thicknesses of less than one inch for the surface coefficient, as follows:

Wind Velocity Miles per Hour	Thickness Inches
0	1.000
1	0.845
2	0.750
3	0.688
4	0.640
5	0.602
10	0.488
15	0.425
25	0.384

The coefficient scale has been drawn for an ambient temperature of 80 degrees Fahr. Since the heat loss from a furnace wall, for ordinary variations in ambient temperature, may be regarded as a function of the difference between surface and ambient temperatures, any other air temperature may be used in calculation by adding a constant amount to the value of each temperature graduation on the surface coefficient scale. For example, if the problem calls for an ambient temperature of 120 degrees Fahr., the 80 degrees Fahr. graduation will be re-labelled 120 degrees Fahr., the 100 degrees Fahr. point will be made 140 degrees Fahr., the 200 degrees Fahr. point 240 degrees Fahr., and so on, and the calculating process will remain unchanged.

Conclusion

In the foregoing pages a graphical method for the solution of problems in steady state heat loss through furnace walls has been developed. In common with all graphical methods, its accuracy is dependent directly upon the size of the chart used, so that the accuracy of the known data may always be matched in accuracy of computation by proper choice of chart size.

The outstanding advantage of the graphical method is that solutions of accuracy sufficient for engineering use may be obtained quickly and surely, avoiding lengthy computations so baffling to one not frequently using them.

DISCUSSION

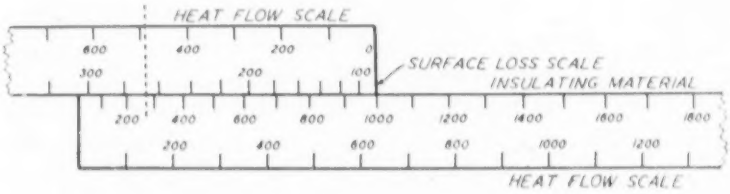
Written Discussion: By J. B. Austin, Research Laboratory, U. S. Steel Corp., Kearny, N. J.

All those who have had occasion to make the tedious calculations involved in estimating the heat loss through a compound wall will welcome the short cuts provided by this ingenious method devised by Mr. Weinland. It may appear to some, and especially to those who are not familiar with such a compu-

tation, that the method is still rather complicated, but if they will repeat the calculation for the sample problem given by Mr. Weinland, using the method now generally employed, the advantage of the new method will become evident.

One of the most interesting features of Mr. Weinland's paper is the use of median thermal resistivities to obtain the first estimate of the heat flow. Thermal resistivity, which is in many cases more easily handled than its reciprocal, conductivity, does not seem to have received the recognition it deserves. In the present case it makes the first approximation a very simple calculation, and gives a result which would be sufficiently accurate for a great many purposes.

In a few special applications, when many calculations involving a limited number of more or less standard thicknesses of material are to be made, the method can be rendered still more convenient. Thus, taking Mr. Weinland's first example (scales 1B and 1C) if we draw the scales on strips of cardboard and invert the surface loss scale over the other, so that the air temperature on the former is directly over the hot face temperature on the latter, the temperature of the cold face will be that at which the two scales coincide. This procedure is illustrated in the following diagram. If the scales are set so that 80 degrees Fahr. on the surface loss scale is above 1000 degrees Fahr. on the scale for the material, the scales coincide close to 270 degrees Fahr. (dotted line in diagram) which is the temperature of the cold face of the insulating



material; and the heat flow can be read directly from the scale on the upper strip—490 B.t.u./hr./sq.ft.

This modification of the method requires, of course, a different scale for each thickness of material, but if many calculations are to be made the convenience justifies the preparation of scales for the few thicknesses which are likely to be used in practice. For example, if it is desired to estimate the heat loss through an uninsulated open-hearth furnace roof, composed of 12 inches of silica brick, for different furnace and ambient air temperatures the cardboard scales are very useful.

For walls composed of several kinds of brick, movable scales of this sort can be used instead of the dividers, but again it is necessary to have a separate scale for each thickness so that unless a number of calculations are to be made it is likely Mr. Weinland's reduced scales will be found the more convenient.

Written Discussion: By R. P. Heuer, Director of Research, General Refractories Co., Philadelphia.

The procedure described by Mr. Weinland is a very convenient one for applying the trial-and-error method of calculating the flow of heat through composite furnace walls. His use of function scales simplifies that problem and the accuracy obtainable is well within the requirements of the problem from an

engineering standpoint. In order to make full use of the method more accurate data are necessary on the thermal conductivity of various refractory materials. Mr. Weinland has included function scales for fireclay and silica brick. All fireclay brick do not have the same thermal conductivity. The conductivity is influenced by the methods of making the brick, the porosity, temperatures of burning, etc. Fortunately the necessary data are now being assembled to make available to furnace engineers accurate information on the thermal conductivity of the common refractories such as various fireclay brick, silica, chrome, magnesite, etc.

Written Discussion: By J. D. Keller, Assistant to Professor Trinks, Carnegie Institute of Technology, Pittsburgh.

This paper is an excellent piece of work, giving evidence of a large amount of thought on the subject, and presenting a method which should be of value to all those who have to calculate heat flow through walls. The great advantage of the method lies in its applicability to composite walls containing more than two kinds of materials.

In his reference to the graphical method developed by Professor Trinks and described in "Industrial Furnaces", Vol. I, Chapter 3, Mr. Weinland seems to be under the impression that the Trinks method is limited to average instead of true conductivities. This impression was doubtless caused by the fact that in the example given in the book for the construction of the curves, the average conductivity was used. This was done for the sake of brevity, and was considered sufficiently accurate for most uses, especially in view of the fact that, at the time when the method was developed, average conductivities were practically the only ones available. For some years past, however, we have used the method with true conductivity values for materials like slag wool, for which the conductivity-temperature line has a considerable curvature. The temperature-heat conduction curve is constructed by the point-by-point, 100 degree interval method. For example, starting at 2600 degrees Fahr., the heat, conducted when the outside of the wall is at T degrees Fahr., is

$$\frac{1}{\text{thickness}} \times [\,(2600 - 2500) \times C_{2550} + (2500 - 2400) \times C_{2450} + (2400 - 2300) \times C_{2350} + \cdots + (\,(T - 100) - T) \times C_{(T - 50)}]$$

the values C_{2550}, C_{2450}, etc., being the true conductivities at 2500, 2450 degrees Fahr., etc. Once the curve has been obtained for any thickness of a given material, curves for other thickness are found with great ease by the method of inverse proportion.

It should also be noted that the Trinks' graphical method can be applied to walls of more than two materials, by using two or more sheets of sufficiently transparent paper. For a wall of three materials, the method is, if anything, simpler than for two materials. The temperature-heat conduction curve (1) for the innermost layer of material in the wall is laid out, sloping downward to the right as in Fig. 75 of "Industrial Furnaces", Vol. I (3rd edition), on the same (stationary) sheet as the surface heat-dissipation curve (3). On top of this is placed a transparent sheet with the curve (2) for the next layer of material, sloping downward to the left, and over this is laid another transparent

sheet with curve (4) for the third or outer layer of material, sloping downward to the right. The transparent sheets are shifted horizontally with respect to each other and to the stationary chart, until curves (2) and (4) intersect on the y-axis, while the intersection of (1) with (2) is on the same vertical line as the intersection of (4) with the surface heat-dissipation curve (3) (see Fig. 75 referred to above); this vertical line indicates directly the heat loss through the composite wall.

On reading Mr. Weinland's paper, the thought occurred to the writer that it might be possible, by the use of some sort of inclined-axis nomogram, to make one scale as in the author's Fig. 2 suffice for each material, and by changing the inclinations of the various scales to correct automatically for different thicknesses. One way to do this would be to place each scale, drawn to such a length that when horizontal it is correct for the smallest commercially-used thickness of the material, on a separate strip of carboard; pivot the strips on a background sheet, and mark the slopes which apply to various other thicknesses of each material. The background sheet on which the strips containing the scales would be pivoted should be of cross-section paper (graph paper) having vertical and horizontal lines close together. For any combination of materials and thicknesses, the user would set each scale to the proper slope, set the dividers to an estimated value, and apply them not to the sloping scale but to its projection on the horizontal, as indicated by the vertical background lines. The same divider setting would be used for all the components of the wall. Except for this difference, the dividers would be used exactly as described by the author in his section III. It would seem that this modification would obviate the tedious calculation of correction factors and resetting of dividers, and would make the method entirely graphical. Possibly an even simpler way of using the inclined scales could be devised.

The writer feels that one point should be emphasized in connection with any computation of heat loss through walls, namely, that in view of (1), variations of at least 30 per cent in the conductivity of different kinds of firebrick, etc., as well as considerable variations from lot to lot in brick and also in insulating materials, and of (2), variations in heat loss due to heat storage when steady temperature conditions have apparently but not actually been reached (which losses are only distinguishable from true conduction loss with difficulty if at all), any values obtained by computation, graphical or otherwise, are not hard and fixed values, but are to be regarded as rather rough approximations only; and that it is especially misleading to attempt to figure savings of one material over another when the difference between them is of the order of 5 or 10 per cent of the total heat loss.

Written Discussion: By F. A. Harvey, Director of Research, Harbison-Walker Refractories Co., Pittsburgh.

The first part of the paper, through the second section, has been the part which I personally have studied most carefully. The impression which I gather from this introductory part is that the preliminary discussion and mathematical analysis are entirely in accordance with fact up to the middle of page 434, when, however, the *increment* of temperature is taken as 100 to 150 degrees Fahr., it seems to me that this is so far from what we regard as a differential increment that the definition of true conductivity no longer holds. In other words, up to

equation 7, it seems to me that everything is logical and exact, but that the term true conductivity should not be used for so large an interval without additional explanation. If the curve showing the relation between heat quantity and temperature per unit area were a straight line, the quantity dq/dt would be a constant, and true conductivity and average conductivity would have the same numerical value. By the amount, however, which the curve departs from a straight line, the definition of true conductivity will be inaccurate. What I am really intending to convey is that I believe there should be a qualitative statement following equation 7 and also that I do not believe you should call your figure "true conductivity".

R. E. Birch and R. S. Moore, of our company, have considerable experience in calculating heat flow through walls of various types and they have compiled the following discussion of the latter part of your paper:

That heat balances of industrial furnaces and methods of reducing heat losses through the refractories are receiving more and more attention is everywhere recognized. One journal which recently published an article concerning the calculation of heat losses, has reported that the request for reprints has far exceeded that of any other individual article. To the accumulating fund of information on this subject, Mr. Weinland's method is an ingenious contribution.

Persons who only infrequently have occasion to calculate heat losses, find that the published methods, whether graphical or arithmetical, involve tedious operations involving trial and error. Although the repeated use of the best of these methods leads to a measure of simplification based on the worker's experience, the operations still remain laborious and any new calculation method is received hopefully.

In considering Mr. Weinland's method, it is of most interest to determine whether or not the method introduces simplifications which will reduce the time required for calculating heat losses. This involves comparisons of an unfamiliar method with others whose manipulation has become familiar through frequent use, and naturally it is difficult to eliminate presumption from such a comparison.

If we first distinguish between the methods of calculation which are predominantly graphical and those which are predominantly arithmetical, the comparison is simplified. In this grouping, Mr. Weinland's method is considered to lead to a graphic solution, although some steps of multiplication are required. It is our feeling that the new method seems simpler by far than other graphical methods with which we are familiar.

The method does not seem to possess the same marked advantage over the best of the predominantly arithmetical methods. As a matter of fact, Mr. Weinland's method would almost certainly require more time. Allowing for the simplification of method (for example, in checking) which would come with experience in its use, the new method is found to contain a greater number of steps than the best arithmetical methods. This conclusion applies particularly for those who frequently make heat loss calculations. Furthermore, the successive shifting between arithmetical and graphical steps in Mr. Weinland's process might increase rather than shorten the time required to arrive at a solution.

The foregoing statements are made with the frequent user in mind. For another group of persons, the beginners and those who seldom make heat loss calculations, the method is recommended for its concise, easy-to-follow steps. As another welcome simplification, Mr. Weinland's method substitutes multiplication for the steps involving division which predominate the arithmetical solutions. This advantage is of no consequence to slide-rule users.

Written Discussion: By C. A. Peterson, Armstrong Cork Co., Lancaster, Pennsylvania.

Mr. Weinland has devised an ingenious method for simplifying the calculations and shortening the time usually required for solving problems on furnace heat losses.

As described, the procedure consists of two parts. (1) The total heat loss, assigning a certain resistivity to each element, computing its resistance, totalling the several resistances and dividing into the temperature difference. (2) The application of these several results modified by a "working factor" to a series of functional scales obtained from a series of laboratory tests on thermal conductivity. The first operation is the conventional method of calculating heat losses by the formula for transmittance which is equal to the reciprocal of the sum of the total resistances of the several elements of the wall. If the calculation is completed in the orthodox manner, the temperatures of the various dividing planes will not be in error as much as 5 per cent with Mr. Weinland's final calculations. Since the materials used in furnace walls will usually vary more than 5 per cent due to manufacturing conditions, it would appear that the old formula gives results within the limits of good engineering practice. The furnace engineer would not design a wall where the hot face temperature was within 5 per cent of failure since the operation of the furnace might cause accidental failure by slightly overheating.

The functional scales necessary for the second part of the calculation are obtained only by a series of expensive tests in apparatus located in very few laboratories in this country. A manufacturer may provide his engineers with functional scales covering his materials but in order to design composite walls where alien materials are specified together with his own, he would have to test competitive materials in the same pains-taking manner in order to complete his data.

Since every manufacturer is constantly pushing development work to improve his products, the original work would be useless in a year both for his own and alien products as well. New materials are constantly appearing on the market which would further increase the burden of tests.

Other physical qualities are equally as important as thermal conductivity such as spalling, expansion, softening under load at high temperature, abrasive resistance and flame cutting resistance, etc., so that undue emphasis should not be laid on thermal conductivity.

Finally, the entire heat distribution in the wall may be upset by the manner in which the furnace is run where the furnace wall is not sealed air tight. I refer to air infiltration, where the furnace may be run either on stack draft with a negative furnace pressure or positive pressure such as usually found with blowers and oil burners. With a negative pressure and no air sealing a very poor insulation may be made to perform extremely well while

with a positive pressure a high quality insulator may perform very badly or actually fail due to the hot side of the insulation being raised beyond its upper limit. In some types of insulation 50 per cent difference in performance can be easily shown in the two cases.

Written Discussion: By P. Nicholls, Supervising Engineer, Fuels Section, U. S. Bureau of Mines, Pittsburgh Experiment Station, Pittsburgh.

The best method of issuing data on heat transmission has frequently been a subject of dispute. Because of the variations of both conductivity and of surface coefficients, tables become cumbersome and it is not possible to restrict them and yet maintain exact agreement with experimental values. Conductivities and surface coefficients can be given more concisely as plots, but there has always been the doubt whether the use of curves would be understood by all, though there should be less objection to this method now than there was ten years ago.

Fig. 2 of the paper suggests a very condensed method and, if accompanied by a sheet on which the various types of computation are reduced to formulae, it should be usable without much study. The writer suggests that three single-figure scale values would be enough, as the accuracy aimed at is not warranted; the surface coefficient varies with conditions, and an error is introduced by using one scale for all conditions; also, even though the conductivities may be fairly accurate for the samples tested, yet for most materials they may vary considerably for different consignments. To be complete, Fig. 2 should have included a scale for common brick.

The writer in a paper[1] in 1922 suggested a graphical method for one or more thicknesses of insulation based on using true conductivity, but he went farther than that adopted by the author in that computation was avoided by the use of alinement charts; thus the trial and error are accomplished by moving straight edges. This method is a great time saver when a large number of computations are required—such as for a publication. Once having made the basic chart, the scales for conductivities or surface coefficients may be replaced by scales for other materials or conditions, and thus any degree of normal accuracy can be maintained.

The alinement chart is convenient in that one can read off the required thickness for a given heat flow or given temperature at the surface, also the thickness for a limiting temperature at the junction between a double layer insulation.

Written Discussion: By Edgar C. Rack, Consulting Engineer, Johns-Manville Sales Corporation, New York.

Mr. Weinland's presentation of a graphic method for the solution of a particular type of problem in heat transfer is concise and clear, and merits careful perusal and consideration. The paper is instructive and should prove of considerable interest to those responsible for the design, operation and maintenance of furnaces.

This paper presents an opportunity for a general discussion on the comparative value of graphical and arithmetical methods for the solution of en-

[1] P. Nicholls, "The Derivation of True Thermal Conductivity Coefficients from Overall Test Results," *Transactions*, American Society of Heating and Ventilating Engineers, Vol. 28, 1922, p. 325-342.

gineering problems. The question of which method is used may be considered largely one of personal ability and inclination. Many engineers prefer the analytical, while others like the geometrical method of dealing with their problems. Some may employ both methods, using one as a check against the other. After all, it would seem that preference of method is largely a matter of the point of view of the user, and for this reason it cannot be stated dogmatically that one or another method is the best, as long as either gives results satisfactory for engineering purposes.

The engineer is primarily interested in the speed and accuracy with which any method can be applied for the solution of his practical problems. In this timely development, successive steps are arranged in logical order; supplementary computations are simplified and "bookkeeping" entries are reduced to a minimum. The facilities for speedy utilization with the added advantage of a satisfactory degree of accuracy of results for engineering and designing purposes is, therefore, evident for the service of those who prefer graphical solution methods.

Formulae are often used with little more than a superficial knowledge of their derivation or limitations. However, in order to set up geometrical figures or function scales which are required for use with graphical method of solution, it should be necessary for the engineer to analyze the formula for which he is to use this "tool" as a substitute. It is also essential for him to make a critical survey of all available data upon which to base the construction of his geometric figures and scales. It may then be said, in favor of the method of solution proposed by the author, that those who adopt it for their purposes will acquire a more exact understanding of the fundamentals of heat transfer, as well as a clearer perception of the thermal qualifications of refractory and insulating materials.

It should be emphasized that any function scale for the surface loss factor must be based upon data of which our present knowledge is still incomplete. A definite need exists for further clarification of the effects of size, shape, and orientation of the surface, as well as its nature and the effect of air movement, upon the rate of heat loss from surface to air. However, in furnace wall design, since the surface resistance is generally low in relation to the wall conduction resistance, a considerable change in the value of the surface loss factor must be made before there is any appreciable resulting change in the total resistance to heat flow. The function scale built up by the author is considered sufficiently accurate for all engineering uses in the field of furnace walls, especially when it is taken into account that a far more important source of error may arise from neglect to consider the effects of permeability and pressure difference between the two sides of the furnace wall. The actual heat loss through a furnace wall may be greater or less than that calculated on the basis of conduction only, depending upon whether there is a higher pressure on the hot or cold side of the wall, and the degree of permeability of the wall[2]. Even after insulating with some of the loose filling materials offered for insulating purposes, the permeability coefficient of the wall may be appreciable. The in-

[2]H. Bansen, "Gas Permeability of Silica Brick, Mortar Joints and Masonry," *Arch. Eisenhüttenwuls*, Vol. 1, 1928-29, p. 687-92. See also E. Maase, "Calculation of Losses from Industrial Furnaces," *Feuerfest-Offenban*, Vol. 7, No. 7, July 1931.

dustry is able, however, to supply materials in preformed shapes which, when properly applied, are capable of accomplishing material reductions in wall leakage. If this method of solution could be amplified to include the factors of pressure difference and wall permeability, it is felt that its use could be more widely extended, for those who prefer graphical methods for the solution of engineering problems.

The author has undoubtedly devoted much time and study to the subject, and the method he is offering, as a labor-saving device for solving problems which involve lengthy computations, should certainly be welcomed by all who have occasion to make heat transfer calculations of furnace walls or similar equipment.

Written Discussion: By R. H. Heilman, senior industrial fellow, Mellon Institute of Industrial Research, Pittsburgh.

A number of graphical methods have been presented from time to time for the determination of heat transmission through compound insulation. This method is apparently the best that has appeared, and, when the scales are accurately prepared and sufficient temperature intervals are included, the results obtained will be sufficiently accurate for all engineering purposes.

The method should be especially valuable to engineers who have a great many calculations to make for various wall thicknesses, temperatures, etc. However, for the average engineer who makes a few calculations at infrequent intervals, it is doubtful if the graphical method will be of any material assistance.

If one has to construct function scales for the numerous other types of heat insulations not included in the paper, it will be found that the calculations of heat transfer can be made much quicker by the trial and error method of mathematical calculation.

Since the graphical method also requires some mathematical calculation as indicated in the third section, it is believed that little time will be saved, even with all function scales at hand, over that of the mathematical method. The simplicity of the mathematical method is indicated in a recent paper by the writer[3]. When the mean temperature conductivity curves for materials not having a straight line relation are obtained from accurate test results with small temperature differences between the faces of the material, no error will result in their use.

Authors' Closure

Many interesting and important points have been brought out in the preceding written discussions, among which the subject that probably deserves the greatest emphasis is that of "true" thermal conductivities. This method of representing the result of thermal conductivity tests is undoubtedly the most accurate one proposed thus far, and the growing interest in the subject is indicated by its further treatment in a recent paper by Finck[4]. That a test run with temperature differences of the order of 150 degrees Fahr. yields data only

[3]Technical Bulletin No. 47, American Refractories Institute.

[4]J. L. Finck, "An Apparatus for Measuring the Thermal Conductivity of Refractories at High Temperatures," *Journal*, American Ceramic Society, Vol 18, 1935, p. 6.

approximately the mathematical concept of true conductivity, as brought out by F. A. Harvey, is quite true. Closer approximations are not generally needed in present day engineering practice, and in any case further developments in testing technique will be required to make them available.

Having obtained the conductivity data for the materials of furnace construction in the form of true conductivity curves, the question then arises as to how to use the data in calculation. To use such data in the conventional arithmetical calculation method, as is apparently advocated in the discussion by R. H. Heilman, involving as it does the assumption of equivalence between mean conductivity integrated over a wide temperature interval and true conductivity at the mean temperature, is simply to perpetuate the type of error which creeps in when conductivities are other than straight line functions. What is needed is some method of integrating the true conductivity function between limits variable in such a manner as to be applicable to any problem. In the discussion by J. D. Keller a practical solution of the problem is presented as an extension of Trinks' method of calculation, which is further developed to enable calculations involving more than two materials in the wall to be made. The use of the adjective "tedious", in the writer's original discussion of Trinks' method, was somewhat ill-advised.

The integration of the k dt function and the representation of the integral as a function scale is a thoroughly workable and accurate device. As brought out in the discussion by P. Nicholls, this device was proposed by him in 1922, and developed into a completely graphical, nomographic method, although its application to calculations involving a multiplicity of materials has only been indicated in a broad way. The suggestion of J. D. Keller of using function scales of variable slope would result in a completely graphical method, and is therefore quite interesting.

In the restricted field of calculations involving one material only and the surface coefficient, or two materials, the suggestion contained in the discussion by J. B. Austin, of inverting one scale, is exceedingly handy, and may be readily extended to take care of a material in various thicknesses by a chart consisting of the function scale for one inch thick material drawn horizontally at the bottom, with lines from each temperature graduation converging to a point near the top of the page, and with other horizontal lines at distances below the point which are inversely proportional to other thicknesses of the material, to any one of which lines the surface coefficient scale may be applied in the inverted manner. Care must be taken that all scales are drawn to the same scale factor in using Austin's modification of the method.

The point stressed principally by R. P. Heuer, and also by P. Nicholls, J. D. Keller, and C. A. Peterson, that the results of calculation are no better than the fundamental conductivity data upon which they are based, is entirely true. The development of conductivity testing in the direction of reduced cost, standardization of method and wider application to commercial materials is going ahead at an accelerated pace at present, but much still remains to be done. Likewise, there is much more information needed regarding the effects of gas flow through the furnace wall, as mentioned by E. C. Rack and C. A. Peterson, and the effect of nonattainment of thermal equilibrium, as mentioned by J. D. Keller.

Regarding the diversity of opinion as to whether this method is generally preferable to the conventional arithmetical method, the suggestion of E. C. Rack, that individual inclination should be the controlling factor, entirely disregards the possibilities of this method for avoiding the errors inherent in the mean-temperature assumption, both in the expresson of conductivity results and in their use in calculation. The suggestion of C. A. Peterson, that the use of median resistivities only is entirely within the necessary accuracy for engineering practice, is believed to be misleading, for it is less accurate than the method generally understood as the "conventional" one, which would be better represented by the one described by R. H. Heilman. The attitude of J. B. Austin regarding the median resistivity method is preferable, because it implies that judgment must be exercised in every case before the method is decided to be sufficient.

The suggestion of R. H. Heilman, that this method will be of greatest use to the frequent user, and that of R. E. Birch and R. S. Moore, that the beginner and infrequent user will benefit most, stand in interesting contrast.

THERMAL EXPANSION CHARACTERISTICS OF SOME NICKEL CAST IRONS

By T. J. Wood

Abstract

The thermal expansion characteristics of some ferritic as well as austenitic nickel cast irons have been investigated, particular attention having been devoted to those compositions which possess generally useful engineering properties.

In the low alloyed, ferritic cast irons, limitations in the nature of less resistance to heat and corrosion narrow the applicability of these compositions to fairly low temperature ranges and but mildly corrosive surroundings.

The non-magnetic austenitic cast irons have high expansion characteristics, and in order to obtain equivalent resistance to heat and corrosion with lower expansivities it is necessary to enter the ranges of about 27 to 45 per cent nickel.

The substitution of nickel plus copper (in the Monel ratio) for an equivalent amount of nickel in cast iron produces expansion rates comparable to the straight nickel alloys at 15 and 20 per cent, but at higher alloy contents the expansivities are of a higher order than in the corresponding copper-free cast irons.

IN RECENT years the greater refinements made in engineering construction have created a demand for cast irons which are not only heat resistant and corrosion resistant, but which also must have thermal expansion characteristics suitable for the particular application involved. However, efforts on the part of the designing engineer to select cast irons of the optimum expansivity have been considerably hampered because the information available on the precise effect of the various "commercial" elements on this physical property of cast iron has been somewhat meagre. It is in an attempt to relieve this situation, partially at least, that the expansion data in the present paper are offered. It is to be emphasized that these data are not presented as a systematic study of the effects of varying nickel con-

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tents on the expansivity of a fixed cast iron base mixture, but rather as an assembly of information on various alloyed cast irons having compositions adjusted to produce other useful engineering properties as well.

Literature

It is well known, of course, that cast iron is essentially a heterogeneous mixture, and that it contains at least six elements, all of which may affect the net expansion rate in some certain degree. Probably one of the most powerful factors affecting the expansivity of ordinary cast iron is the combined carbon content. For example, it has been shown by Driessen[1] that an increase in the combined carbon content from 0.05 to 3.66 per cent in iron will decrease the mean coefficient by as much as 26 per cent at 212 degrees Fahr., and 17 per cent at 750 degrees Fahr. (400 degrees Cent.). Silicon promotes the formation of ferrite (decreasing the combined carbon content) and in this function it may be expected to increase the expansion rate (i.e., over that of the pearlitic or sorbitic irons) although the element per se does not appear to greatly affect the expansivity of iron, as indicated by Greiner, Marsh and Stoughton[2]. Available data[3] on the coefficient of expansion of graphite show that the presence of this element in its free form tends to decrease the coefficient very slightly.

The similarity in the expansion coefficients of the austenitic nickel-copper-chromium cast iron known as Ni-Resist, and certain cast aluminum alloys has already been taken advantage of commercially, as recorded by Jeffries[4], as well as Vanick and Merica[5]. In this instance, Ni-Resist sleeves were used in aluminum alloy cylinders, particularly in air-cooled engines and light weight air compressors.

Boegehold[6] reported that the thermal expansion coefficient of an "adjusted base" 1.41 per cent nickel cast iron is appreciably lower than that of the usual grade of plain cast iron. His results also indicated that additions of molybdenum tend to lower the coefficient below that of the plain iron.

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[1]Driessen, International Critical Tables, Vol. II, p. 470, Table 19.

[2]E. S. Greiner, T. S. Marsh and Bradley Stoughton, "The Alloys of Iron and Silicon," Alloys of Iron Research Monograph Series, McGraw-Hill Book Company, 1933.

[3]International Critical Tables, Vol. II.

[4]Zay Jeffries, "Light Metals in the Automotive Industry," American Society for Testing Materials, Symposium on Developments in Automotive Materials, 1930, p. 64.

[5]J. S. Vanick and P. D. Merica, "Corrosion and Heat Resistant Nickel-Copper-Chromium Cast Iron," Transactions, American Society for Steel Treating, Vol. 18, 1930, p. 923.

[6]A. L. Boegehold, "Present Day Methods in Production and Utilization of Automotive Cast Iron," American Society for Testing Materials, Symposium on Developments in Automotive Materials, 1930, p. 5.

A detailed study of the expansion characteristics of cast irons ranging in nickel content from 20 per cent up to 50 per cent has been presented in an article by Wickenden[7], in which he mentions the use of Ni-Resist for piston rings, and also for valve seat inserts in aluminum engine heads. This author showed that austenitic cast irons of high heat resistance and corrosion resistance could be obtained with expansion characteristics which very closely match those of aluminum, plain cast iron or carbon steel, depending upon the nickel content.

Apparatus

The apparatus used in determining the expansivity of the various cast irons was the quartz-differential dilatometer which has been described by Scott[8]. The details and limitations of this apparatus are too well known to require any discussion here.

The thermal gradient of the heating furnace in the test zone was practically zero, due to a copper jacket around the outer quartz tube of the dilatometer. The heating rate of the test specimens averaged 7.3 degrees Fahr. per minute. A dial gage which could be read in units of 1/10,000 of an inch was used to measure the length changes of the specimens.

Test Specimens

All of the cast irons used in this work were cast in the form of bars ¾ inch in diameter by 6 inches long, with the exception of one white cast iron which was ¼ inch in diameter by 6 inches long in the cast dimensions.

Approximately three-quarters of all of the specimens were machined or ground with the ends tapered at about a 45-degree angle with an effective length of about 4 inches, and ¼ inch in diameter. Following the Scott procedure it was necessary to take into consideration the error introduced in the linear expansion measurements by the lateral expansion of the conical ends of the specimens, inasmuch as the quartz seats for these ends had a much lower expansion rate. Due to an improvement in the dilatometer design, the remaining quarter of the specimens were tested with flat ends, and had an effective length of about 4 inches with a diameter of ¼ inch. The thermocouple was attached by means of a force fit of the bead in a

[7]T. H. Wickenden, "Heat and Corrosion Resistant Cast Irons," *Automotive Industries*, Vol. 69, 1933, No. 15, p. 428.

[8]Howard Scott, "Expansion Characteristics of Low-Expansion Nickel Steels," Transactions, American Society for Steel Treating, Vol. 13, 1928, p. 829.

Table I

Analyses and Calculated Expansion Coefficients of Test Specimens

Specimen	Reference Figure	Reference Curve	Analysis						**Effective Length of Specimen L	**Correction Factor	Mean Coefficient of Expansion, in Millionths Per Degree Fahr., from 50°F. to Indicated Temp.						
No.	No.	No.	TC	Si	Ni	Cr	Cu	Mn	(Inches)	K	200	300	400	500	600	700	800
2470	2	1	3.15	2.16	.....	....	.....	.42	4.000	1.000	6.68	6.80	7.09	7.24	7.39	7.45	7.50
7238	2	2	2.98	1.23	2.24	....	.....	.50*	3.880	0.968	6.53	6.69	6.78	6.89	7.01	7.09	7.12
7296	2	3	2.84	0.87	3.30	....	.....	.44	4.001	1.000	6.30	6.40	6.65	6.90	7.21	7.34	7.45
7295	2	4	2.84*	1.32	3.30*	....	.....	.44*	3.984	1.000	6.96	7.06	7.10	7.24	7.33	7.43	7.52
7297	2	5	2.84*	1.89	3.30*	....	.....	.44*	3.999	1.000	6.62	6.90	7.12	7.30	7.44	7.49	7.57
7239	3	1	2.98*	1.23*	7.38	....	.....	.52	3.900	0.968	7.25	7.25	7.39	7.58	7.66	7.86	7.42
7240	3	2	2.98*	1.23*	12.58	....	.....	.50*	3.880	0.968	9.45	9.78	9.86	9.90	9.91	9.86	....
7241	3	3	2.98*	1.23*	20.56	1.61	.....	.50*	3.890	0.968	9.58	10.15	10.39	10.36	10.34	10.28	....
2776	3	4	3.15*	1.75*	13.96	2.13	6.00*	1.00*	3.880	0.968	10.28	10.68	10.70	10.67	10.65	10.66	....
7385	.	.	2.36	1.36	25.74	1.96	.....	1.00*	4.000	1.000	6.96	8.00	8.59	8.91	9.12	9.22	....
W-84	4	1	2.46	1.55	29.20	....	.....	.90	3.910	0.968	4.26	5.05	5.96	6.57	7.05	7.38	7.62
W-85	4	2	2.57	1.60	29.56	1.22	.....	1.20	3.910	0.968	4.09	4.95	5.67	6.29	6.82	7.19	7.46
W-87	4	3	2.71	1.59	30.96	2.64	.....	.93	3.910	0.968	3.43	4.36	5.25	5.91	6.42	6.85	7.20
W-86	4	4	2.57	1.54	31.60	1.21	.....	1.16	3.905	0.968	3.60	4.27	4.97	5.70	6.30	6.74	7.04
2220	4	5	2.50	1.27	34.50	3.59	.....	1.00*	3.900	0.968	3.11	3.48	3.99	4.49	5.04	5.53	5.96
7242	4	6	2.50*	1.25*	34.70	2.60	.....	1.00*	3.890	0.968	2.29	2.89	3.50	4.23	4.96	5.50	6.00
W-91	5	1	2.25	1.65	43.68	3.21	.....	1.80	3.905	0.968	4.76	5.06	5.32	5.48	5.62	5.86	6.14
W-96	5	2	2.35	1.46	44.80	3.97	.....	.97	3.895	0.968	4.77	4.97	5.13	5.26	5.40	5.78	6.00
W-97	5	3	2.46	1.40	46.45	3.93	.....	.94	3.900	0.968	4.60	4.77	5.05	5.32	5.54	5.68	5.99
W-95	5	4	2.13	1.36	49.96	....	.....	.27	3.915	0.968	4.42	4.95	5.32	5.43	5.47	5.44	5.54
6800	5	5	1.90*	1.40*	60.00*	4.00*	.....	1.00*	4.000	1.000	7.13	7.40	7.51	7.63	7.57	7.61	7.97
7223	5	6	1.70*	1.40*	70.00*	4.00*	.....	1.00*	4.000	1.000	7.11	7.50	7.73	7.80	7.94	8.03	8.13
W-133	6	1	2.60	1.00*	20.38	2.47	9.08	1.00*	3.900	0.968	7.75	8.64	8.96	9.14	9.24	9.32	....
W-105	6	2	2.27	1.53	30.82	4.19	15.10	.96	3.900	0.968	4.77	5.17	5.48	5.82	6.17	6.49	6.82
W-109	6	3	2.15	1.46	34.84	4.03	17.03	1.01	3.830	0.920	5.58	5.96	6.06	6.22	6.42	6.62	6.80
2506	6	4	2.95	.90	4.86	1.55	.....	.40*	3.930	0.973	4.32	4.83	5.46	5.96	6.29	6.62	6.87
8089	.	.	2.85*	1.50*	10.70*	2.00*	4.30*	1.00*	4.000	1.000	9.67	10.08	10.32	10.46	10.55	10.59	....
8090	.	.	2.00*	1.50*	28.20*	4.00*	11.30*	1.00*	4.001	1.000	4.66	5.24	5.78	6.29	6.64	7.06	....

*Approximate

**Mean Coefficient $= \left(\frac{dL}{T_f - T_o} \times \frac{K}{L} \right) + Q$

where dL = increase in length of specimen
T_f = final temperature
T_o = original temperature
K = (1.0 — correction for lateral expansion of conical ended specimens) where $K = 1.0$, flat ended specimens were used.
L = effective length of specimen
Q = coefficient of expansion of Vitreosil = .0000003 (per °F.)

small hole drilled in the center of the specimen. The analyses of the test specimens are given in Table I.

Test Results

The expansion curves obtained on the several cast irons are shown in Figs. 1 to 6. A slight departure from the orthodox method was employed in calculating the expansion coefficients in that the slope of the lower portion of any given expansion curve was extended

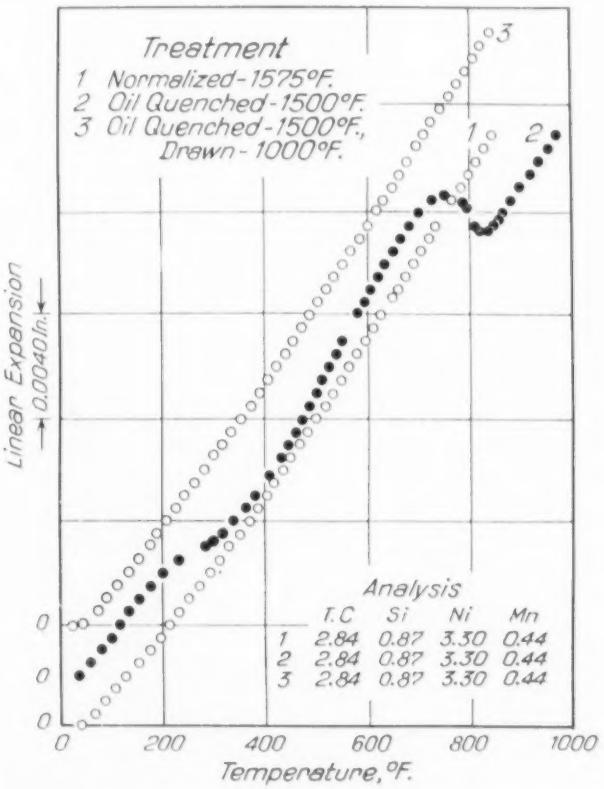

Fig. 1—Expansion Curves of 3.30 Per Cent Nickel Iron After Treatment Shown.

until it met the 50 degree Fahr. intercept. By noting the corresponding displacement of the zero point by this extension, a correction was made for the length change of the specimen for each temperature selected in the calculation of the mean expansion coefficients. This procedure was felt to be required when it was considered that the lower portion of practically all expansion curves (secured by the "continuous reading" differential method) are bowed toward the left, an effect due to the thermal lag between the thermocouple and

the specimen during the initial stages of heating. Consequently, by extrapolating the general slope of the lower portion of the curve to some arbitrarily selected minimum temperature, a more nearly accurate expansion curve should be obtained. Of course, this method may be fraught with danger in those cases where low temperature

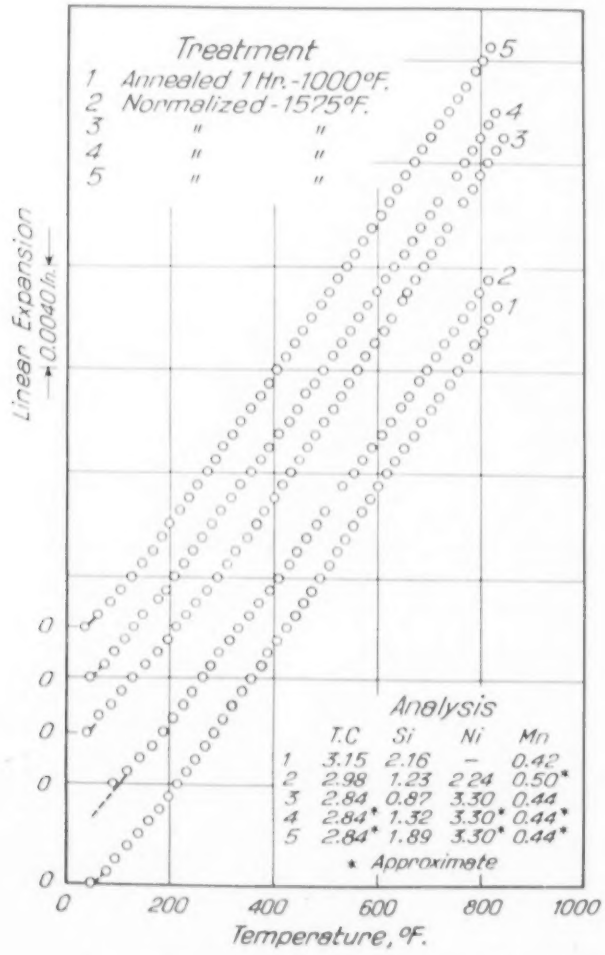

Fig. 2—Expansion Curves of Iron Having Constant Nickel Content (3.30 Per Cent) and Varying the Silicon Content. Heat Treatment as Shown.

transformations occur, but the absence of such phenomena in the present work made the step permissible. It may be noted that in several instances where curves were obtained starting at temperatures well below the base point of 50 degrees Fahr., the corrections became extremely slight, and under no circumstances were they very large.

All of the cast iron expansion specimens were annealed prior to testing, as it had been found that the length changes due to casting

strain relief on the first run were of a magnitude sufficient to prevent reproducible results.

Evidence of the variables which may be initially encountered in the expansivity of plain or low alloyed cast iron due merely to differ-

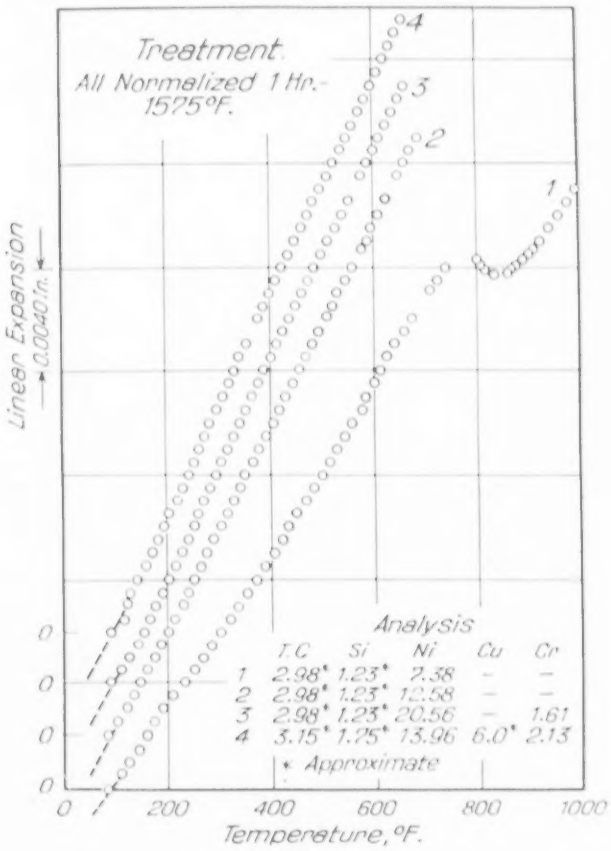

Fig. 3—Expansion Curves of Martensitic Cast Irons After Heat Treatments Shown.

ences in cooling rates is contained in Fig. 1, which shows the expansion curves of a nickel cast iron which has been:

(1) Normalized

(2) Normalized, and then oil-quenched from 1500 degrees Fahr. (815 degrees Cent.)

(3) Treated as above in (2), and drawn at 1000 degrees Fahr. (540 degrees Cent.).

It will be noted that the expansion rates of this iron in conditions (1) and (3) are practically identical, whereas in the hardened condition the expansion curve bears little resemblance to its mates.

The effect of an increasing silicon content (with the nickel held

constant) upon the expansivity of cast iron is demonstrated by Curves 3, 4 and 5 in Fig. 2. The expansion coefficients of these irons are shown in Table I, and demonstrate, in general, that as the silicon content increases, the coefficient increases. The trend is not

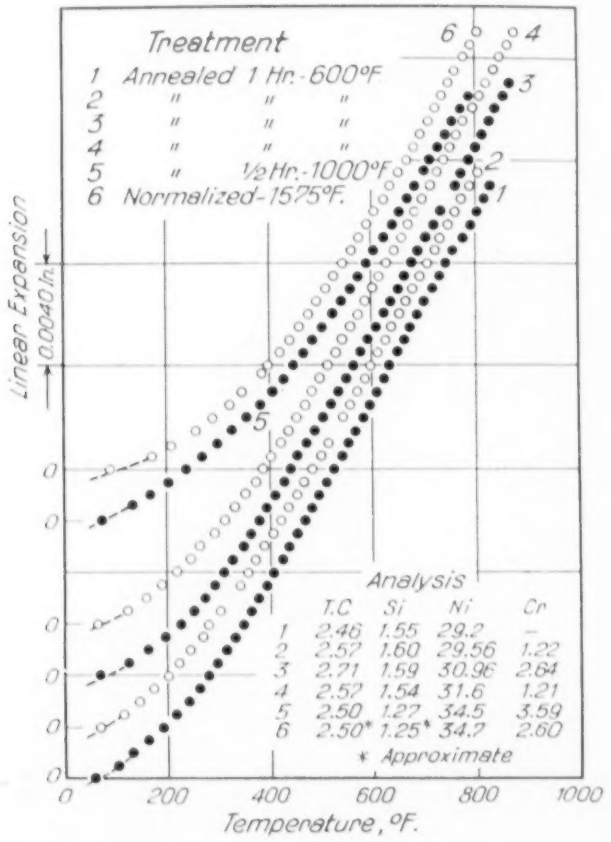

Fig. 4—Expansion Curves of Austenitic Cast Irons with Varying Nickel and Chromium Contents After Heat Treatments Shown.

clearly defined at temperatures below 400 degrees Fahr. (205 degrees Cent.), but it is well established in the higher ranges.

The expansion characteristics of martensitic and austenitic cast irons are exemplified by Curves 1 to 4 in Fig. 3. The break in Curve 1 is due, of course, to the transformation from the ferritic to the austenitic condition. The compositions represented by Curves 3 and 4 are the usual types of austenitic cast iron, and are structurally stable over a wide temperature range. The chromium content of these irons is present for the purpose of increasing the hardness and also to enhance the thermal stability.

In Fig. 4 and 5 are given the expansion curves of cast irons ranging in nickel content from 29.2 per cent up to about 70 per cent. The majority of these irons contain substantial amounts of chromium in order to provide a cast hardness of around 140 to 160 Brinell. They are easily machinable.

In view of the similarity in the expansion coefficients of the austenitic 20 per cent nickel and the 14 per cent nickel-6 per cent

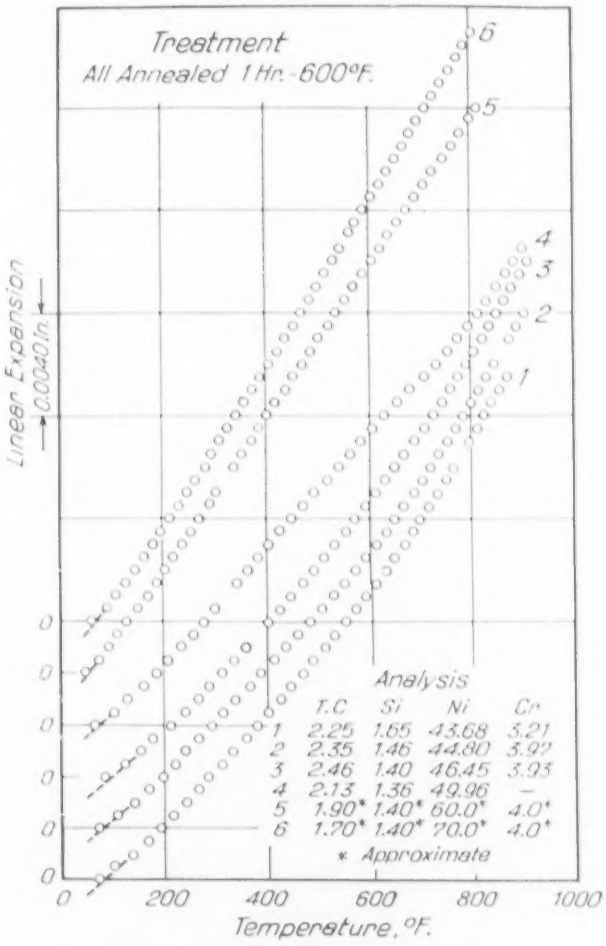

Fig. 5—Expansion Curves of Austenitic Cast Irons with Varying Nickel and Chromium Contents After Heat Treatments Shown.

copper cast irons (see Curves 3 and 4, Fig. 3), it was felt desirable to ascertain whether the substitution of nickel plus copper (in the Monel ratio) for an equivalent amount of nickel would yield identical expansion rates in other alloy ranges. The expansion curves of three of the compositions selected are represented by Curves 1, 2 and 3

of Fig. 6. The mean coefficient data of these irons may be found in Table I along with the values obtained for a 15 per cent Monel and a 39.5 per cent Monel cast iron. Analysis of these data in Table I failed to reveal any close relationship in the expansion characteristics between the high nickel plus copper (or Monel) irons and the nickel

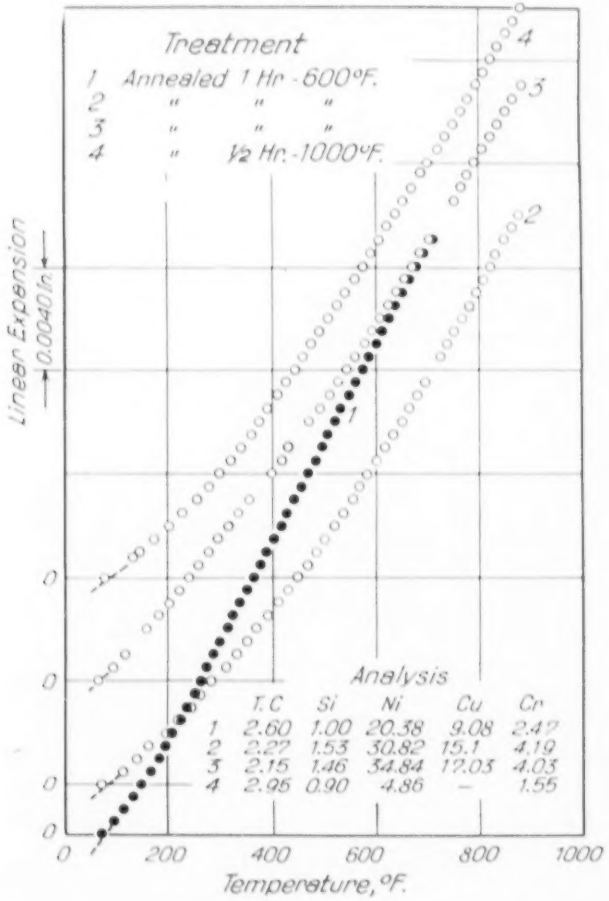

Fig. 6—Expansion Curves of Austenitic Cast Irons with Varying Nickel, Copper and Chromium Contents After Heat Treatments Shown.

irons of equivalent alloy content, whether considered as total equivalence (nickel plus copper), as nickel alone, or as a change in the iron-nickel ratio with the copper acting as a diluent. Thus it may be concluded that the substitution of nickel plus copper in the Monel ratio for an equivalent amount of nickel in order to obtain similar expansion characteristics seems to be efficacious only in the lower ranges, as at 15 per cent Monel (10.7 per cent nickel plus 4.3 per cent copper), or 20 per cent Monel, while at higher alloy ranges the cop-

per-containing irons possess higher expansion rates than the corresponding copper-free compositions.

A comparatively low expansion coefficient is obtained in the hard, white, alloyed cast iron represented by Curve 4 in Fig. 6. The fact

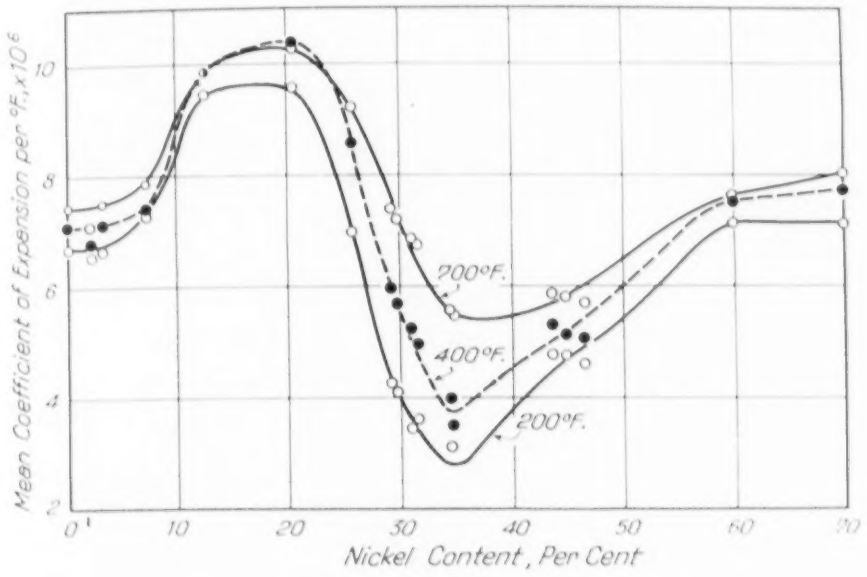

Fig. 7—Expansion Isotherms of Nickel Cast Irons.

that all of the carbon is in the combined form accounts for the low expansion rate of this iron.

The isotherms shown in Fig. 7, which represent the general trends which may be expected at the various nickel ranges in cast iron, do not include the data on the copper-content irons. No attempt has been made to locate these curves with exactitude, because, as mentioned previously, variables such as silicon, carbon, and manganese contents will alter the values somewhat.

In Fig. 8 are shown the isotherms for the nickel-copper cast irons. A comparison of these curves with those in Fig. 7 clearly reveals the higher expansion rates of the nickel-copper irons when the alloy content is in excess of 20 per cent.

By strictly observing compositional limits it is thus possible to obtain heat and corrosion resistant austenitic cast irons which offer a rather wide selection of expansion ranges. As an example, one firm is now using 35 per cent nickel cast iron dies in the manufacture of paper pulp products, as this particular iron furnishes the necessary resistance to heat and corrosion as well as a low expansion rate which

Table II

Typical Properties of Several Austenitic Cast Irons

Heat No.	Type of Cast Iron	Analysis						Ultimate Tensile Strength in P.S.I.	BHN	Usual Range of Ultimate Tensile Strength P.S.I.	Usual Range of BHN	Usual Range of Transverse Strength 12" Span 1¼" Arb. Bar Pounds
		TC	Si	Ni	Cu	Cr	Mn					
2366	Ni-Resist	3.18	1.50	14.62	6.60	2.59	1.0*	26,700	160	20/35,000	130/170	2500/4500
2350	{Cu-free {Ni-Resist}	3.18	1.00	21.28	....	2.63	1.0*	24,500	137	20/35,000	130/170	2500/4500
2406	30% Ni	2.87	1.30	28.60	....	1.71	1.0*	18,750	140	15/25,000	130/160	2000/3500
2263	"Invar"	2.61	2.05	36.07	....	3.71	1.0*	16,160	127	15/25,000	120/150	2000/3500

*Approximate.

prevents excessive distortion in the electrically heated, conical shaped female die. Another concern has found that by substituting 28 per cent nickel cast iron cylinder liners for those made of plain iron in pumps handling gas high in H_2S content, expansion characteristics

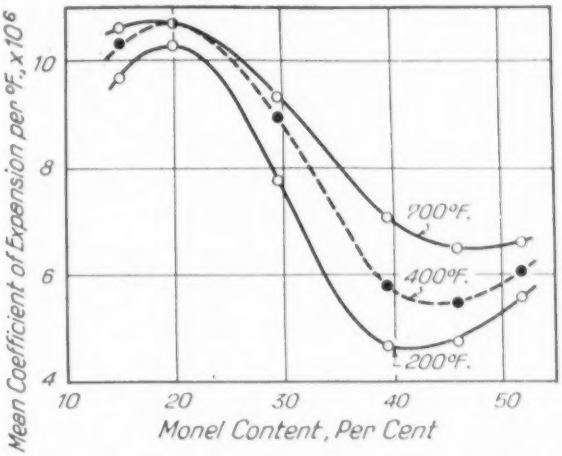

Fig. 8—Expansion Isotherms of Monel (Nickel plus Copper) Cast Irons.

similar to the plain iron liners were obtained, while the lives of the pumps were greatly extended due to the improvement in corrosion resistance. Ni-Resist has been found to be an excellent material for use as an internal chiller core[9] in cylindrical castings of plain iron, as the austenitic cast iron expands rapidly when the molten metal first contacts it so that when solidification is complete, and cooling takes place, the more rapidly contracting Ni-Resist core may be easily withdrawn from the plain iron casting. In a similar manner, numerous other applications and combinations can be made to obtain conformity of expansion properties together with satisfactory heat and corrosion resistance.

Physical Properties

In view of the fact that the physical properties of the low alloyed ferritic cast irons are generally well known, no attempt will be made to review them here. The less familiar properties of several types of austenitic cast irons are shown in Table II, which lists not only the actual values obtained on the different compositions, but also the usual range of properties which may be expected. The principal factors which will influence the tensile strength and hardness of these alloys are the total carbon and chromium contents.

[9] "High Expansivity Feature of Ni-Resist Recommends it for Internal Chillers," *Nickel Cast Iron News*, Vol. 5, 1934, No. 2.

Summary

The expansion characteristics of a series of cast irons ranging in nickel content from 0 to 70 per cent have been investigated. It has been established that many possibilities are afforded in selecting an iron which possesses, besides special corrosion resistant properties, an expansion rate ideal for the application involved (cylinder liners, piston inserts, internal chillers, etc.).

In the low alloyed irons, i. e., the ferritic as opposed to the austenitic types, limitations in the nature of less resistance to heat and corrosion narrow the applicability of these compositions to fairly low temperature ranges and but mildly corrosive surroundings. One exception in this group is the 4.5 per cent nickel-1.5 per cent chromium white cast iron, which has rather remarkable heat resistant properties.

The nonmagnetic austenitic cast irons (18-22 per cent nickel or Monel metal) have high expansion characteristics, and in order to obtain equivalent resistance to heat and corrosion with lower expansivities it is necessary to enter the ranges of about 27 to 45 per cent nickel. The lowest expansion rates of any of the irons investigated were obtained in the 34-35 per cent nickel compositions.

The substitution of nickel plus copper in the Monel ratio for an equivalent amount of nickel in cast iron produces expansion rates comparable to the straight nickel alloys at 15 and 20 per cent, but at higher ranges the expansivities are of a higher order than in the corresponding copper-free cast irons.

Acknowledgment

It is with pleasure that the author acknowledges the suggestions and encouragement given by T. H. Wickenden and N. B. Pilling during the course of this investigation. The skillful assistance of F. E. Kasch in securing the data is also appreciated.

PRACTICAL OBSERVATIONS OF SOME HIGH CARBON HIGH CHROMIUM TOOL STEELS

By W. H. Wills

Abstract

This paper deals with a study of three high carbon high chromium tool steels. One was a straight chromium type, the second resembling the first but with the addition of 1 per cent vanadium, and the third an air hardening type. The points discussed include the manufacture, heat treatment, and physical properties. Special tests comparing the dimensional changes of steels 2 and 3 are described.

Mention is made of the principal applications of these steels and that it is difficult to meet all requirements with a single steel of this type. All things considered, Steel 3 appears to be the most suitable for general uses where a high carbon high chromium steel is desired.

THE high carbon high chromium tool steels as a group are not new, but represent a comparatively recent development among the products of tool steel mills. The high resistance of these steels to abrasion together with good nondeforming properties have made them increasingly popular for many applications. These steels appear to have had their origin from experiments to develop a substitute for high speed steel. They have, however, never found much use in the cutting tool field. They have instead been used in many kinds of die applications and have successfully competed with high speed steel. The first steel of this type on the market had a range of 1.50 to 2.50 per cent carbon and 12 to 16 per cent chromium with no other alloys. This straight chromium type, particularly when the carbon content is over 2.00 per cent, is quite low in toughness or impact value, and for the purpose of improving this, later analyses have come out with additions of vanadium and nickel. Other additions have included cobalt and molybdenum, the latter in particular to improve the air hardening property.

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Gill[1], in his paper on high carbon high chromium steels, covers in a comprehensive way the metallography and physical properties of six representative types of these steels. In view of their widespread use it was thought that more information regarding these steels, particularly of a practical nature would be of quite general interest. The purpose of this paper, therefore, is to discuss three types of high carbon high chromium steels with which the author has been concerned involving their manufacture and application. These data are based on mill observations, laboratory tests and examination of tools that have been in service. The first of these is of the older straight chromium type while the other two contain a similar chromium content and the addition of other alloys. The type analyses of these steels are as follows and they will be designated by these numbers throughout the paper.

	Steel No. 1 Per Cent	Steel No .2 Per Cent	Steel No. 3 Per Cent
Carbon	2.40	2.10	1.55
Manganese	.35	.30	.30
Silicon	.30	.25	.30
Chromium	11.50	12.00	12.00
Vanadium	....	1.00	.25
Molybdenum	....	....	.80

Melting

The melting of these steels can be done in the crucible or by the electric furnace process. At the present time it is carried out for the most part in electric arc furnaces with a basic hearth.

A typical electric furnace charge consists of about 30 per cent high carbon high chromium scrap and the balance high grade low phosphorus low sulphur steel scrap such as heavy plate or punchings. When melted, the first preliminary test will show a carbon content of approximately 75 per cent of the carbon content in the original charge and the recovery of the chromium will run about 80 per cent. The heat is slagged off as usual and carburized on the bare metal with crushed electrodes, resulting in a gain in carbon of approximately 0.70 to 0.80 per cent. The bath is then covered with a mixture of lime, fluorspar and crushed coke to make up the deoxidizing slag. After a white slag is obtained and the temperature of the bath increased somewhat, the addition of ferrrochromium is made—in sev-

[1]J. P. Gill, "High Carbon High Chromium Steels," TRANSACTIONS, American Society for Steel Treating, Vol. XV, No. 3, p. 387.

eral parts—as the amount to be added is comparatively large and if too large a quantity is added at once the bath and slag will be chilled excessively.

Additions of other alloys of course, are relatively quite small compared to the ferrochromium and these are made a short time before the heat is ready for pouring. The length of time under the deoxidizing slag is increased due to the large ferrochromium addition and the total time averages 30 to 40 minutes longer than on a heat of high speed steel.

The pouring temperature of these steels is comparatively low and is quite close to that of high speed steel. The straight chromium steel (Steel 1) is more fluid then the others and Steel 2 with the high vanadium content appears the most sluggish. All three grades are susceptible to piping and must be poured slowly. The common sizes of ingots poured are 8-inch square (700 pounds) and 12-inch square (1500 pounds).

Forging, Rolling, Annealing

Ingots of all three of these grades are always annealed before forging. Slow and thorough heating before cogging is very important. Ingots and the larger billets (4-inch square or larger) are preheated before charging in the heating furnaces. The usual practice after a day's run is to open the door on the heating furnaces and allow them to cool down to about 1400 degrees Fahr. before the ingots or billets are charged for working the following day. The charge is allowed to equalize for several hours before the temperature is gradually brought up to about 1900 degrees Fahr. A temperature of 1950 degrees Fahr. should not be exceeded in the case of any of these steels. The heating time on the 8-inch ingots runs about 16 hours in the furnace before hammering and 50 per cent longer on 12-inch ingots. While the heating time is much the same as high speed steel, the forging temperature is 50 to 100 degrees Fahr. lower. Eight-inch ingots are reduced to 4-inch square billets on the hammer and require two reheatings. Owing to the density and brittleness of these steels in the as-cast condition, rolling of the ingots on a mill is not practicable and the carbide network or cellular ingot structure is best broken down by hammering. These steels should not be worked too cold (below approximately 1550 to 1600 degrees Fahr.) because tearing is apt to occur on the corners. After forging or rolling, the stock should cool down out of contact

with air currents or moisture. When the steels are worked into billet form the ductility is increased so that rolling of sections of ordinary size can be accomplished without difficulty. The largest sections commonly rolled are 3.5 to 4-inch rounds. It is difficult to avoid a more or less "woody" fracture associated with a tendency toward cellular carbide distribution in the larger sizes of bar stock such as 3.5-inch round or larger. This means low resistance to shock particularly in a transverse direction. Improved strength is obtained by the use of forgings when these larger sections are involved. Of the three steels discussed, Steel 2 has the narrowest forging range and is the most liable to give trouble with tearing; Steel 3 has a wider range and is the most readily handled in forging and rolling, while Steel 1 is a close second. The per cent recovery from ingots to finished bars will vary somewhat, but 65 to 70 per cent is considered good.

Regular practice at the mill is to anneal the high carbon high chromium steels with high speed steel heats at 1600 to 1650 degrees Fahr. with a cycle of approximately 6 to 8 hours coming up 8 to 12 hours at temperature and 24 to 36 hours cooling. Hardness on annealed bar stock following this cycle will range from 217 to 248 Brinell. While the great majority of users find these steels to machine satisfactorily at these hardness values, there are occasional complaints in respect to their machinability. Their inherent quality for abrasion resistance makes for more or less difficulty in certain machining operations such as milling and filing. This disadvantage is in most cases offset by the advantage of longer tool life.

Cold Drawing

The high carbon high chromium steels as a group are quite susceptible to work hardening and this combined with their brittle nature makes them very difficult to cold draw. The same drafts are used as with high speed steel, but even with these very light reductions they are harder to draw than high speed steel. As might be expected, Steel 3 with the lowest carbon content can be handled the most readily and has been drawn in small flats and rounds. Steels 1 and 2 can be drawn in rounds but it is not very practical to draw sections with sharp corners.

Heat Treatment

These steels with their low heat conductivity absorb heat quite

slowly and require considerable time at or approaching the hardening temperature for proper solution of the carbides. This is necessary for the development of maximum hardness and wearing quality. While the heat treating methods followed by users of these steels vary, the practice of pack hardening is quite generally used. Packing materials may be charcoal, carburizing compound, or cast iron chips. It is possible to acquire a carburized case in pack hardening if fresh charcoal or carburizing compound is used and when this is undesirable, spent materials should be used.

When the heating for hardening is done in an open fire furnace, the atmosphere should be so adjusted that scaling is held to a minimum during the comparatively long heating period. In ordinary electric or fuel-fired furnaces, scaling may be retarded by putting charcoal on the hearth or bottom and avoiding the introduction of air as much as possible. Electric furnaces with controlled atmospheres are now available and have solved the problem of scaling and decarburization encountered in the open fire hardening of these steels, thereby saving considerable time over pack hardening.

On account of the varying conditions that obtain in heat treating furnaces—such as rate of heat transfer, furnace atmosphere, etc.—it is impossible to make a rule as to definite length of time required in hardening steels of this type. However, the curves shown in Fig. 1 will be of interest as indicating the approximate time required to bring Steel 3 up to the hardening temperature in a gas-fired furnace. This is typical for the group and assumes the use of a preheat at 1100 degrees Fahr. The time periods indicated by curve 1 are nearer to that required in the open fire.

A comparison of the hardening temperature of the three steels shows Steel 1 with a range 1650 to 1750 degrees Fahr. and Steel 2 and 3 with 1750 to 1850 degrees Fahr. for developing maximum hardness and grain refinement. With temperatures exceeding 1900 degrees Fahr. grain growth sets in, hardness decreases and accompanying dimensional changes increase the internal stresses and liability of cracking after quenching. Within the preferred hardening ranges, however, the steels have been held for periods up to 3 hours without appreciable grain growth or excessive distortion. Steel 2 with the high vanadium content is more sluggish toward structural changes than the other two and requires a longer time at temperature to produce maximum hardness.

Steels 1 and 2 require oil quenching to develop the hardness

necessary for usual tool and die purposes, while Steel 3 will harden in oil or air. When cooling in still air the parts should be brought out of the furnace as free from scale as possible and the hardness will be almost up to that obtained with an oil quench. To obtain the least distortion after treatment, the tools should always be cooled in still air.

As the high carbon high chromium steels are inherently brittle, it is desirable to draw promptly, particularly following an oil quench-

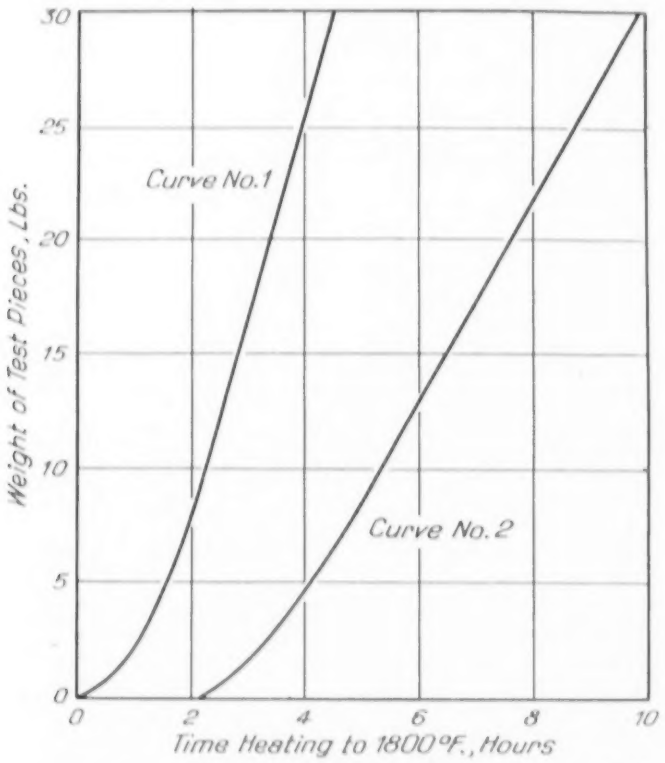


Fig. 1—Curve Showing Time Required to Heat Steel 3. Curve No. 1. Test Pieces Preheated at 1100 Degrees Fahr. Transferred to Charcoal Pack at 1800 Degrees Fahr. Curve No. 2. The Test Pieces were Packed in Charcoal and Brought Up with the Furnace to 1800 Degrees Fahr for 1 Hour 30 Minutes.

ing operation. Steel 1 with the lowest impact value is the most likely to develop cracks from hardening strains and the lowest draw that will properly relieve these, unless the size is quite small, is about 300 degrees Fahr. The most commonly used drawing temperatures range from 400 to 600 degrees Fahr. and to a less extent up to 1000 degrees Fahr. With slow heat penetration, long drawing treatments are the rule and some users make it a practice of allowing the charge to cool with the drawing furnace.

When the hardening temperature of these steels has been on the high side so that the structure as-quenched is largely austenitic it is possible to develop secondary hardness with drawing treatments at approximately 900 to 1000 degrees Fahr. This is more pronounced with higher hardening temperatures and it is possible to thus obtain an increase in hardness from as low as Rockwell C 42 to 45 (charac-

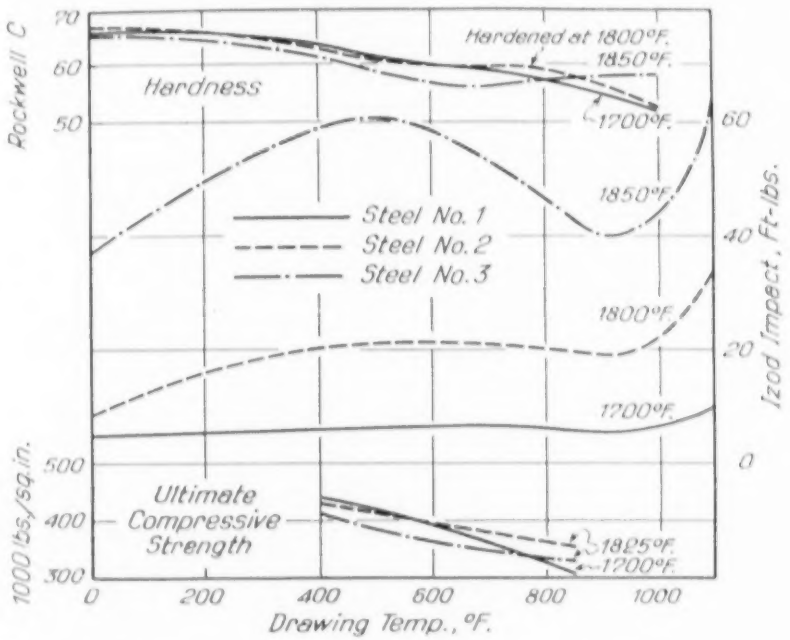


Fig. 2—Variation of Hardness with Drawing Temperatures.
Fig. 3—Impact Versus Drawing Temperatures.
Fig. 4—Ultimate Compressive Strength Versus Drawing Temperature.

teristic of the fully austenitized structure) up to as high as Rockwell C 60 to 62. So far as is known no particular advantage is gained by such a treatment and increased dimensional changes would result in a greater possibility of tool breakage.

Physical Properties

Numerous tests have been made on these steels in the laboratory to study their physical properties. The description and results that follow are limited to those that were considered of most importance to the tool maker. There are included variation of (1) hardness, (2) impact or toughness value, (3) compressive strength, (4) dimensional changes after hardening at standard temperatures followed by different drawing treatments.

For the drawing treatment series test pieces were turned 1-inch in diameter from 1⅛-inch round and cut 3 inches long. Those representing Steel 1 were heated at 1700 degrees Fahr. and quenched in oil, those of Steel 2 were heated at 1800 degrees Fahr. and quenched in oil and those of Steel 3 were heated at 1850 degrees Fahr. and cooled in air. They were then drawn for 1 hour at the various temperatures. These results were as follows: (see Fig. 2)

Table I
Variation of Hardness with Drawing Temperatures

Drawing Temperature Degrees Fahr.	Rockwell C Hardness		
	Steel 1	Steel 2	Steel 3
As quenched	66.0	66.5	65.5
300	65.3	64.0	64.5
400	61.0	61.5	62.5
500	60.7	60.5	59.0
600	60.3	59.5	56.5
700	58.5	59.0	56.0
800	57.0	59.0	57.0
900	55.3	57.0	58.0
1000	51.3	51.4	58.0

Steel 1 was hardened from 1700 degrees Fahr.
Steel 2 was hardened from 1800 degrees Fahr.
Steel 3 was hardened from 1850 degrees Fahr.

For the impact series unnotched test pieces were machined from ½″ square bar stock to usual size .394″ square. These were hardened at the same temperatures as for the series used for the hardness tests and drawn 1 hour at temperature in each case. The impact tests were made on an Izod machine and are the average of 5 tests for each heat treatment.

These are shown in Fig. 3 and Table II.

Table II
Impact Properties in Relation to Drawing Temperature

Drawing Temperature Degrees Fahr.	Impact Values in Foot-Pounds		
	Steel 1	Steel 2	Steel 3
As quenched	4.6	8.9	37.0
400	8.0	20.0	56.0
500	5.7	21.3	63.8
600	5.5	21.1	59.2
700	6.1	19.5	52.4
800	7.1	22.5	46.4
900	5.0	18.7	....
1000	5.8	21.8	42.0
1100	9.3	33.3	63.0

Steel 1 was hardened from 1700 degrees Fahr.
Steel 2 was hardened from 1800 degrees Fahr.
Steel 3 was hardened from 1850 degrees Fahr.

For determining compressive strength test pieces were turned 0.505 inch in diameter about 0.015 inch full and cut 1.5 inches long. They were then hardened—Steel 1 at 1700 degrees Fahr. and Steels 2 and 3 at 1825 degrees Fahr. quenching in oil. All specimens were drawn one hour at temperature in each case and then ground to size (0.505 inch). The values for ultimate compressive strength are the average from two test pieces run on a 100,000-pound universal testing machine (see Fig. 4 and Table III).

Table III

Ultimate Compressive Strength in Relation to Drawing Temperature

Drawing Temperature Degrees Fahr.	Compressive Strength in Pounds per Square Inch		
	Steel 1	Steel 2	Steel 3
400	438,500	425,500	403,625
600	396,625	389,250	364,750
850	302,125	350,375	329,000

Steel 1 hardened from 1700 degrees Fahr.
Steel 2 hardened from 1825 degrees Fahr.
Steel 3 hardened from 1825 degrees Fahr.

The specimens for the expansion tests were cylinders 1 inch in diameter and 4 inches long with the ends ground parallel. The length of each was measured with a micrometer. They were then pack hardened—Steel 1 at 1700 degrees Fahr., Steel 2 and Steel 3 at 1800 degrees Fahr., holding 1 hour at temperature. Steel 1 and Steel 2 were quenched in oil and Steel 3 was cooled in still air. The length was measured after hardening and after each draw—cumulative draws one hour at temperature. These results reduced to linear expansion per inch are shown in Table IV and Fig. 5.

Table IV

Linear Expansion in Relation to Drawing Temperature

Drawing Temperature Degrees Fahr.	Expansion per Linear Inch		
	Steel 1	Steel 2	Steel 3
As quenched	+0.00085	+0.00142	+0.00087
400	+0.00063	+0.00110	+0.00050
500	.....	+0.00080	+0.00037
600	+0.00061	.....	.....
750	.....	+0.00037	−0.00007
800	+0.00013	.....	.....
1000	−0.00030	+0.00042	+0.00050

Steel 1 hardened from 1700 degrees Fahr.
Steel 2 hardened from 1800 degrees Fahr.
Steel 3 hardened from 1800 degrees Fahr.

The results of the foregoing tests show that Steel 3 develops more secondary hardness than the others with drawing treatments at 900 to 1000 degrees Fahr. Steel 1 and Steel 2 maintain similar hardness values as the drawing temperature is increased. The impact tests show that Steel 3 is decidedly tougher than the other steels. This

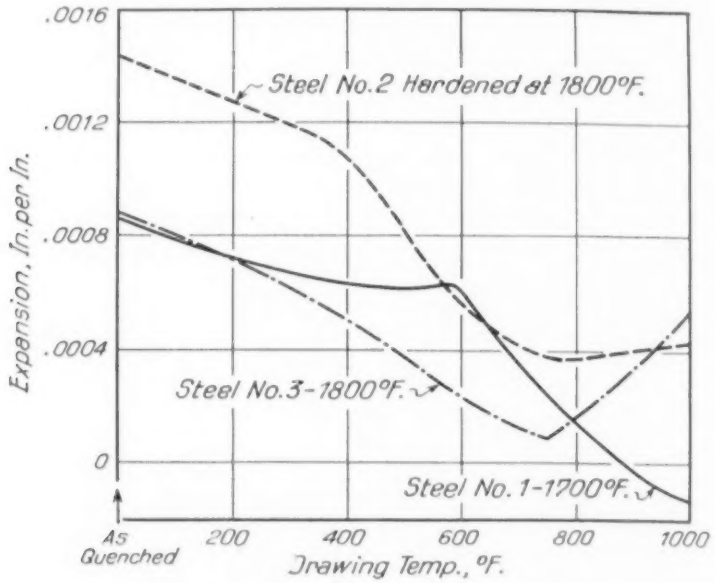


Fig. 5—Expansion Versus Drawing Temperature.

is probably due to a lower carbon content. Steel 1 with the highest carbon content is the most brittle. It developed the maximum compressive strength with the lower draws but this decreased considerably with the 850 degrees Fahr. draw. Steels 2 and 3 showed little difference in the compressive strength developed. Steel 2 showed more change in length than Steels 1 and 3 which were similar with the lower draws.

Special Tests

A series of tests were run to compare Steels 2 and 3 as to change of size after various heat treatments and time held at temperature. All test pieces were turned 7/8 inch round from 1-inch round taken from stock. These were cut 4 inches long and the ends ground parallel. The length was checked with a micrometer at center punch marks on end of test pieces.

After length measurements were made the test pieces were subjected to the following heat treatment:

Preheated in an electric furnace at 1100 degrees Fahr. When up to temperature, they were transferred to a tube partially filled with charcoal at hardening temperature. They were held in the charcoal pack one hour. The Steel 3 pieces were removed and cooled in still air, while Steel 2 pieces were quenched in oil. Samples of the two grades representing each hardening temperature were run together. These temperatures ranged from 1700 to 1950 degrees Fahr. in steps of 50 degrees.

The size of each specimen was measured in the as-quenched condition, hardness checked and the sample taken for micro examination. Samples of each grade representing each hardening temperature were drawn at 400, 500, 750 and 1000 degrees Fahr. (cumulative draws) and checked for hardness and size between draws. To determine effect of time at temperature on change in size, additional samples were run at 1700, 1750, 1800 and 1850 degrees Fahr. in the same way, but held 3 hours at temperature. Results obtained are given in Tables V and VI. See Figs. 6 and 7.

Table V
Microstructures Observed on Quenched Specimens

Hardening Temperature Degrees Fahr.	Steel 2 Structure	Steel 3 Structure
1700—1 hr.	Largely fine martensite	Fine mixed martensite and austenite
1750—1 hr.	Largely fine martensite	Fine mixed martensite and austenite
1800—1 hr.	Largely fine martensite	Fine mixed martensite and austenite
1850—1 hr.	Martensite and austenite, slight trace of grain boundaries	Mostly austenitic—grain boundaries visible
1900—1 hr.	Martensite and austenite, slight trace of grain boundaries	Mostly austenitic—grain boundaries visible
1950—1 hr.	Martensite and austenite, considerably enlarged needles	Austenitic—rather large grain size
1700—3 hr.	Mixed martensite and austenite, no grain boundaries	Fine mixed martensite and austenite
1750—3 hr.	Mixed martensite and austenite, no grain boundaries	Fine mixed martensite and austenite, trace of grain boundaries
1800—3 hr.	Mixed structure, martensite and austenite, no grain boundaries	Fine mixed martensite and austenite, trace of grain boundaries
1850—3 hr.	Mixed martensite and austenite, trace of grain boundaries	Mostly austenitic. Grain boundaries pronounced

The results of these tests show that Steel 3 has a hardening range of 1800 to 1900 degrees Fahr. wherein maximum hardness is developed—the hardness being over Rockwell C 65—and at the same time maintaining low distortion. This range for Steel 2 is 1750 to 1900 degrees Fahr. for developing a hardness of Rockwell C 65 or higher. However, the change of size for Steel 2 is some-

what greater than for Steel 3. Where the average distortion for Steel 3 over the range 1800 to 1900 degrees Fahr. is an expansion of 0.0009 inch per inch of length as cooled in still air, Steel 2 over the range 1750 to 1900 degrees Fahr. shows an average expansion of 0.00152 inch per inch of length as quenched in oil. After a 2-hour

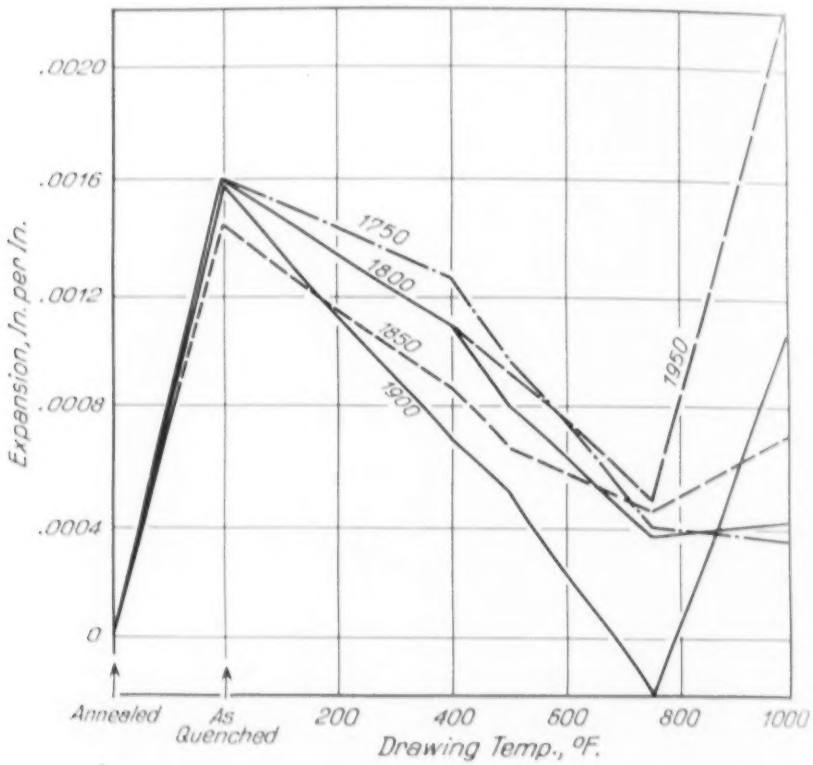


Fig. 6—Change Per Inch of Length with Various Heat Treatments of Steel 2.

draw at 400 degrees Fahr. to Rockwell C 62-64 which is common for many tools, the average change on Steel 3 was 0.00052 inch expansion per inch and on Steel 2 0.0010 inch expansion per inch. Holding a long time at temperature (3 hours) over the range 1700 to 1850 degrees Fahr. appears to have little effect on hardness, structure or dimensional change for either grade.

Throughout the hardening range investigated (1700 to 1950 degrees Fahr.) with both grades there is expansion after hardening. This is followed by gradual contraction with draws up to 750 to 800 degrees Fahr. Above 800 degrees Fahr. there is a pronounced expansion due to structural change from austenite to martensite or troostite.

Table VI
Effect of Heat Treatment on Dimensional Changes of Specimens of Steel 2 and Steel 3 as to Distortion in Hardening

Test Number		Heat Treatment	Original Length	As Quenched Change	As Quenched R.H. C	400 Draw Change	400 Draw R.H. C	500 Draw Change	500 Draw R.H. C	750 Draw Change	750 Draw R.H. C	1000 Draw Change	1000 Draw R.H. C
Steel 2	1	1700 1 hr. Oil	4.0244	+.0038	64.0	+.0029	61.5	+.0022	60.0	+.0009	58.0	—.0004	52.0
Steel 2	2	1750 1 hr. Oil	3.9929	+.0064	68.0	+.0051	63.5	+.0035	61.5	+.0016	60.0	+.0014	54.0
Steel 2	3	1800 1 hr. Oil	3.9918	+.0062	68.5	+.0044	64.5	+.0032	62.0	+.0015	60.5	+.0017	54.0
Steel 2	4	1850 1 hr. Oil	4.0012	+.0058	68.0	+.0035	64.0	+.0026	62.5	+.0018	60.5	+.0028	58.0
Steel 2	5	1700 3 hr. Oil	4.0058	+.0065	68.0	+.0042	65.5	+.0037	64.0	+.0016	62.0	+.0032	56.0
Steel 2	6	1750 3 hr. Oil	4.0053	+.0061	68.0	+.0040	65.0	+.0033	63.0	+.0013	61.0	+.0022	56.5
Steel 2	7	1800 3 hr. Oil	4.0260	+.0060	67.0	+.0038	64.0	+.0030	63.0	+.0010	61.5	+.0029	56.5
Steel 2	8	1850 3 hr. Oil	4.0018	+.0058	67.5	+.0033	63.5	+.0027	62.5	+.0007	61.0	+.0040	57.5
Steel 2	1-A	1900 1 hr. Oil	4.0144	+.0059	65.5	+.0027	63.5	+.0021	62.0	—.0008	61.0	+.0042	60.5
Steel 2	2-A	1950 1 hr. Oil	3.9834	+.0062	52.5	+.0044	52.0	+.0037	52.5	+.0019	52.0	+.0096	63.0
Steel 3	1	1700 1 hr. Air	3.9990	+.0040	64.0	+.0035	61.5	+.0029	61.0	+.0015	59.0	+.0020	54.0
Steel 3	2	1750 1 hr. Air	3.9895	+.0034	63.0	+.0025	61.0	+.0024	60.0	+.0011	58.0	+.0018	56.0
Steel 3	3	1800 1 hr. Air	4.0120	+.0035	66.5	+.0020	63.0	+.0015	62.5	—.0003	61.0	+.0020	59.0
Steel 3	4	1850 1 hr. Air	4.0060	+.0035	65.5	+.0016	62.0	+.0006	62.0	—.0007	60.0	+.0040	63.0
Steel 3	5	1700 3 hr. Air	3.9872	+.0035	65.5	+.0018	62.5	+.0009	62.0	—.0005	60.5	+.0022	59.0
Steel 3	6	1750 3 hr. Air	4.0010	+.0035	67.0	+.0013	63.5	+.0008	62.0	—.0015	60.0	+.0020	60.0
Steel 3	7	1800 3 hr. Air	4.0034	+.0039	65.0	+.0019	62.0	+.0013	61.5	.000	60.0	+.0039	61.0
Steel 3	8	1850 3 hr. Air	3.9698	+.0032	65.0	+.0012	62.0	—.0003	61.5	—.0016	60.0	+.0042	63.0
Steel 3	1-A	1900 1 hr. Air	3.9836	+.0038	65.0	+.0026	62.0	+.0015	61.0	+.0007	60.5	+.0050	62.0
Steel 3	2-A	1950 1 hr. Air	3.9792	+.0040	56.0	+.0025	54.0	+.0022	53.0	+.0020	53.0	+.0082	59.5

Table VII
Effect of Air Hardening and Oil Hardening on Dimensional Change of Specimens of Steel No. 3

Test Number	Heat Treatment	Original Length	As Quenched Change	As Quenched R.H. C	375 Draw Change	375 Draw R.H. C	500 Draw Change	500 Draw R.H. C	750 Draw Change	750 Draw R.H. C	1000 Draw Change	1000 Draw R.H. C
Set 1	1850 20 min. Air	4.0022	+.0029	67–68	+.0024	65	+.0018	63	+.0005	60.5	+.0034	62
Set 1	1850 1 hr. Air	3.9986	+.0028	65.5–67	+.0025	64.5	+.0010	61.5	—.0008	59.5	+.0030	61.5
Set 1	1850 3 hr. Air	3.9990	+.0027	64–66	+.0027	63	+.0012	60.5	—.0002	59	+.0041	63
Set 2	1850 20 min. Oil	4.0032	+.0049	67–68	+.0010	64.5	+.0022	62	+.0010	59.5	+.0033	61
Set 2	1850 1 hr. Oil	4.0035	+.0047	66–67	+.0041	64.5	+.0026	61.5	+.0009	59	+.0044	62
Set 2	1850 3 hr. Oil	4.0190	+.0062	63.5–65	+.0058	62	+.0028	60.5	+.0019	59	+.0048	63

When hardening temperatures of 1900 degrees Fahr. are exceeded the hardness falls off due to formation of a larger proportion of austenite in the structure. Dimensional changes are more pronounced, particularly with the higher draws and this is a likely cause of tool breakage, especially with large sections. From a structural

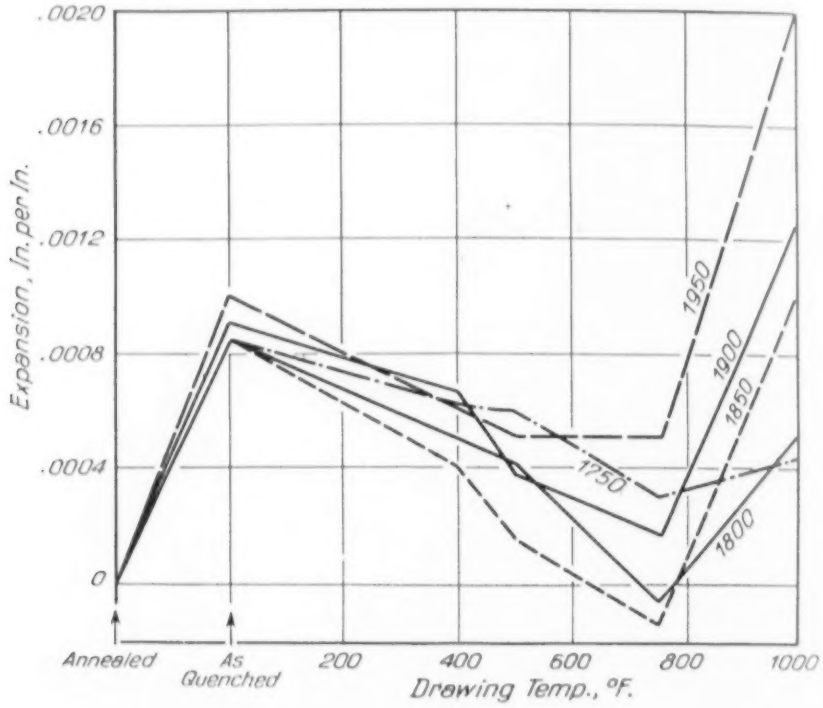


Fig. 7—Change Per Inch of Length with Various Heat Treatments of Steel 3.

standpoint, Steel 3 appears to develop an austenitic condition more readily than Steel 2. In other words, Steel 2 appears a little more sluggish toward structural changes likely due to the higher vanadium content.

Another series of tests were run to compare the distortion of Steel 3 when hardened by cooling in still air and when quenched in oil. Test pieces were machined to size 1.75 by .75 by 5.5 inches long. They were then drilled on the center line for the insertion of ¼-inch dowel pins. The dimension measured with micrometer calipers was outside to outside of the dowel pins. The measurement was made before and after hardening and after each draw. All test pieces were preheated in an electric furnace to 1100 degrees Fahr. allowing about 1 hour.

They were then transferred to the charcoal pack in a tube at 1850 degrees Fahr. allowing in each case 1.5 hours for the pieces to rise from 1100 to 1850 degrees Fahr. Set 1 consisted of three similar test pieces, was held respectively 20 minutes, one hour, and three hours at a temperature of 1850 degrees Fahr., removed and allowed to cool in still air. Set 2 was handled in the same way, but after removing was quenched in oil. After measurement the pieces were drawn in an electric furnace for 2 hours each time—cumulative draws at 375, 500, 750 and 1000 degrees Fahr. checking the dimension after each. Table VII gives the results. These results show that the distortion as-quenched in oil is considerably more than when cooled in still air, although the drawing treatments at 500 degrees Fahr. and above tend to reduce the difference somewhat.

Applications of these Steels

The applications to which these high carbon high chromium steels have been made are worthy of more than passing notice on account of their variety and the lower costs made possible through the high rate of production obtained. Their use for blanking and cold forming punches and dies is one of the most important. In lamination punches and dies they have given phenomenal runs as compared to straight carbon or manganese oil hardening steels on blanking light-gage silicon sheets. By proper tool design they have been successfully used in punching steel plate up to ¼-inch thick. Very long runs have been obtained on cold trimmer dies in drop forge and bolt and nut plants. The lower carbon air-hardening type has also been used for the hot trimming of forgings.

Steels 2 and 3 have given good results on shear blades and slitting cutters used on light-gage material such as sheet, tin plate and strip and even for shear blades on spring steel. Steel 2 has done well on cold forming rolls used on rim stock and seamless tubing, also on seamer rolls for steel barrels and other containers which applications require high resistance to abrasion and high compressive strength.

Steel 2 finds considerable use for drawing dies on steel bars and to some extent for the drawing of stampings. On the former application the steel is not used in the fully hardened condition. The so-called punch dies are put in service in the semi hard condition and when worn are peened on the reverse side, re-punched to size cold and are ready for another run. Sectional dies are made up from

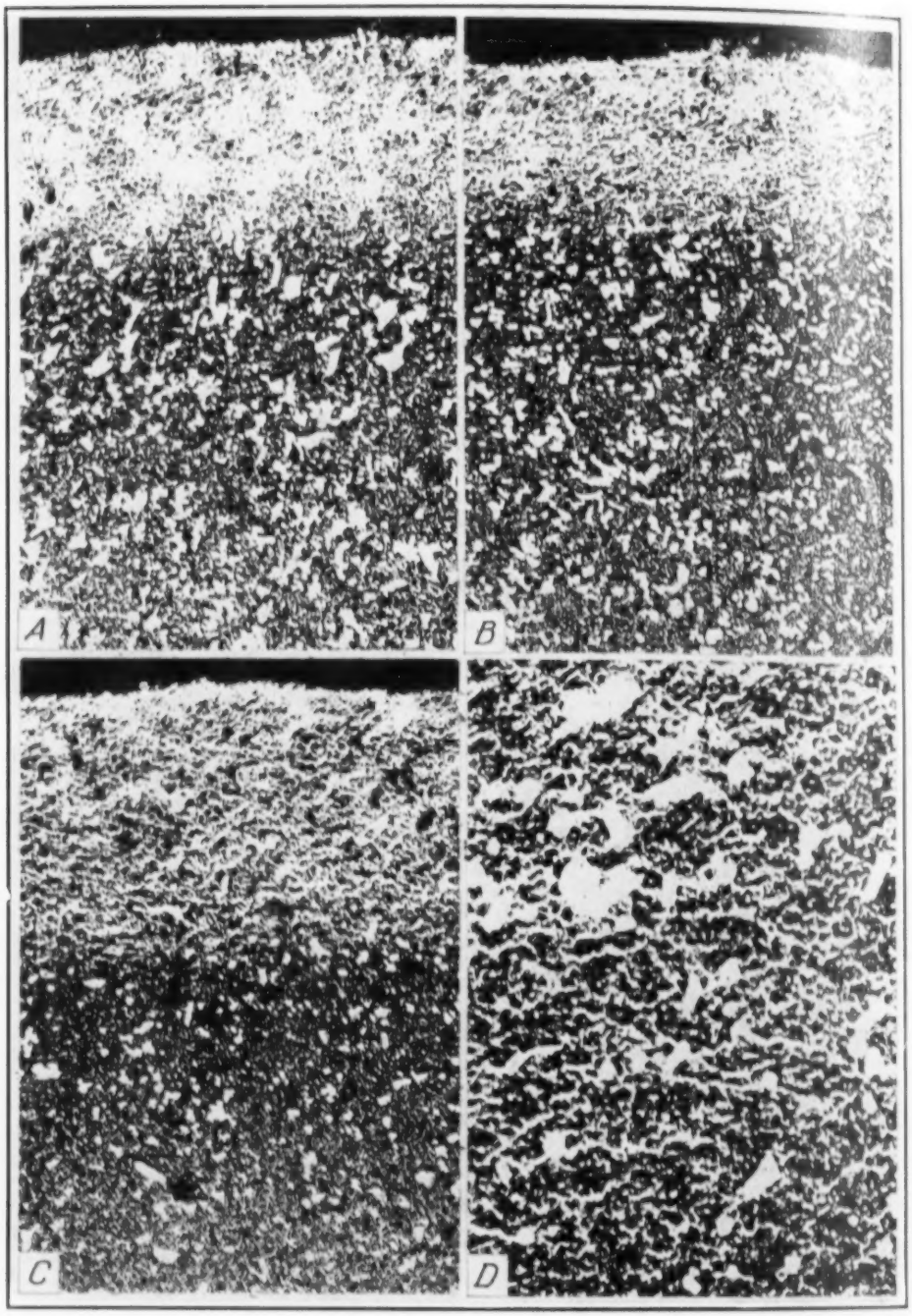


Fig. 8—Nitrided Cases of the Three Steels After the 90-Hour Cycle at 950 Degrees Fahr. A—Nitrided Case of Steel 1; B—Nitrided Case of Steel 2; C—Nitrided Case of Steel 3. All × 200. D—Nitrided Case of Steel 3 Showing Peculiar Network Structure of Exterior Portion. × 600.

semi-annealed stock and put in service without hardening and redressing is done by filing.

The three types have been used for thread rolling dies on long runs and particularly for the larger sizes and coarser threads. Very small dies for fine threads are generally of straight carbon or one of the manganese oil hardening steels. Distortion is sometimes a problem on the larger thread rolling dies and to get away from this, the air hardening type like Steel 3 has been the most popular. Another point in favor of it is that it machines better than the higher carbon steels such as Steels 1 and 2 which means less wear on the threading hobs. In cutting the threads the practice in some shops on steels of this type is to use a speed 30 per cent less than is used on high speed and manganese oil-hardening steels. Steel 3 has given good production on the more difficult thread rolling jobs such as on some of the S. A. E. alloy and stainless steels.

Steel 3 will take a deeper nitrided case than the other two—a case depth of 0.012 to 0.014 inch may be expected following a 90-hour nitriding cycle. See Fig. 8. The other two steels run at the same time developed a case only about one half as deep. Steel 3 hardened in the usual way, drawn at 1000 degrees Fahr. and nitrided has been used for seamer rolls on light-gage material with a large increase in production over high carbon high chromium steels hardened in the usual way. However, it has not worked out successfully on heavy stock where much greater pressures are involved. This grade nitrided has also given much larger runs in certain dies used in the production of collapsible tubes.

Minor applications for these steels include burnishing dies, and rollers, mandrels, liners and dies for forming clay or porcelain, gages and certain kinds of broaches. About the only kinds of cutting tools where they are used to a limited extent are for brass and bronze forming and seating tools.

Conclusion

While at first thought it might appear that a single grade of high carbon high chromium steel would be all that is necessary for a mill to carry as a regular grade, it is apparently impossible to satisfy all users with their varied requirements. Good production records have been made by all three of the steels described, but some will insist that one grade is better than another in resisting wear.

Steel 2 has proved a more popular grade than Steel 1 due largely

to its better resistance to shock, other qualities being much the same. Steel 3 with its air hardening property, low distortion and better machining quality than the others is in good demand, although some claim Steel 2 to be somewhat better as to resistance to abrasion. Taking into account all factors, Steel 3 appears to be the most universally adaptable of the three high carbon high chromium steels.

Acknowledgments

The author wishes to express his appreciation to F. B. Lounsberry, vice-president of the Ludlum Steel Co., for permission to publish information given in this paper. He is also grateful for suggestions regarding melting, forging, and rolling of the high carbon high chromium steels from J. N. Brookfield of the melting department and C. W. Spittal, general superintendent. Credit is due L. R. Lounsberry of the metallurgical laboratory for his part in conducting numerous tests on these steels and compiling the physical test data.

DISCUSSION

Written Discussion: By John E. Erb, General Electric Co., Schenectady, New York.

The information given in this excellent paper is of primary interest to the tool designer. It represents just about the minimum which ought to be furnished him by the steel manufacturer for every brand of steel for dies. In fact it would not be going too far to ask for data on the effect of section on the properties mentioned.

The results of tests and observations we have made agree in general with those of Mr. Wills. However, above all else it must be emphasized that data of this nature apply only to materials of specific composition melted and manufactured by a particular method. Attempts to apply such data to steels of slightly different composition, or even of the same analysis but differently made, will often be found impractical if not misleading.

From the consumer's standpoint it would be quite desirable if all requirements could be met with a single steel of this type. Unfortunately this desire has not been realized so far. This is mainly because of the complexity of the problem and the difficulty of evaluating the results of service tests. So far as our observations go, it appears that the higher carbon variety is best for high production lamination dies for shearing thin silicon sheet. For drawing and forming dies and for porcelain insulator molds and certain types of extruding dies for clay products, the advantage seems to lie with certain brands of the lower carbon grade. However, the results are somewhat contradictory and definite conclusions may not be drawn at this time. We have had trouble

with both classes when used for shearing ¼ to ⅜-inch plate stock. This trouble to some extent may be due to carburization of the surface during pack hardening.

Referring to Tables I and II, I would like to ask Mr. Wills whether the specimens were ground previous to the determination of hardness and impact values? Also, were the impact specimens subjected to hardness tests and, if so, what were the results? In a published report of this nature, I also feel that individual values as well as averages should be included.

On the whole this report will prove most helpful to the tool steel consumers and I can only express the hope that more of a similar type will be forthcoming.

Written Discussion: By A. J. Scheid, Jr., Columbia Tool Steel Co., Chicago Heights, Illinois.

We are very glad to note that Mr. Wills has given considerable thought to what we consider a most interesting field, namely, high carbon-high chromium tool steels. Mr. Wills has obviously done some careful work and has pointed out some important points in connection with the handling of this type of ferrous material.

It has been our thought that steels of the high carbon-high chromium type are worthy of more consideration than has been given them. There is some hesitancy to further utilize steel of this type because of first cost and the somewhat increased difficulty of machining. The ability of these steels to develop high hardness values on air or oil quenching, with considerable retained hardness values up to 900 degrees Fahr. (480 degrees Cent.) on tempering, and their nondeforming properties and high resistance to abrasion, make them particularly useful for a wide variety of applications.

It is to be regretted that Mr. Wills confined his experiments to three compositions. With some thirty-seven different steels of almost as many different analyses offered by American manufacturers, a wider consideration is necessary to indicate the potential physical values of the high carbon-high chromium group.

This is particularly true of his impact tests. The values published are very interesting. The high impact values of steel #3 are due to the decreased carbon content, this decrease in carbon materially reducing the abrasion resistance of the steel. However, further impact values with variation of quenching temperature and time at the quenching temperature are necessary if the impact test is to be utilized as a means of judging the desirability of composition.

In many cases we concur with Mr. Wills, particularly on the fact that the straight chromium type, or a steel with only one modifying element, such as nickel, has a very limited field of application and usefulness and that it takes added counterbalancing alloys to fully develop the major characteristics of the high carbon and high chromium combination.

Our experience has led us to decide also that one composition can give as good results by either oil or air hardening. Our work led us away from the vanadium or vanadium-molybdenum combination to a tungsten-silicon alloy.

We cannot quite agree with Mr. Wills' results in regard to the heat treatment, particularly in regard to the hardening temperatures. It would appear that with the proper combination of time and temperature a slightly lower quenching temperature is desirable and necessary to develop the best physical

properties, including hardness, which, in turn, affects both the tool life as well as the size, shape change and growth in the heat treatment.

Heat treatment perhaps accounts for different results that we have found on similar materials as indicated in Fig. 5 on page 478 of Mr. Wills' paper. This heat treatment difference also probably accounts for the different results in regard to the impact properties as shown in Table II.

These are minor differences only and we offer them for their constructive value in the hope that they will aid Mr. Wills in his further studies of this interesting material.

Written Discussion: By John Lyman Cox, The Midvale Company, Philadelphia.

In his valuable paper Mr. Wills has covered the manufacture of high carbon-high chromium tool steel and given much information on the properties of three varieties of these special steels which are proving useful in many industries. It will be interesting to trace the development of this material.

The first idea of high carbon-high chromium tool steels was probably due to the late P. R. Kuehnrich, of Darwin and Milner, Sheffield. Interested in the manufacture of high grade files, in 1895 he was given in Germany a piece of superior steel which contained 2.5 per cent chromium,—about the figure that Brustlein was using in armor piercing shells. Although Kuehnrich thought this probably too brittle, he made a knife of carbon 1.3 per cent, chromium 3.5 per cent which was excellent. Between 1895 and 1903 Darwin and Milner sold quantities of steel with about 3.5 per cent chromium as files and in other forms.[1] Then in 1903 Kuehnrich decided to go up high in chromium, melting a series of heats with 6, 9, 15, 18, 21 and 24 per cent chromium, and carbon about 1.75 to 2.5 per cent. Apparently he found the best results toward the upper end of his range, for Darwin and Milner then made a draw plate steel known as Durex, with very high carbon and chromium, and their grade known as Neor containing roughly carbon 2 to 2.5 per cent, and chromium 18 to 20 per cent.

A tool maker named Horstman with a shop in the rue St. Maur, Paris, went into the business of casting dies from this material bought from Darwin and Milner, which he called T.C.K. Two Midvale engineers, sent to Paris for a course in the Sorbonne, heard there of a remarkable steel high in chromium, which they reported on their return. Thus in 1908 Midvale began experimenting with such a steel, melting then a heat with carbon 2.5 per cent, chromium 14 per cent. Various analyses were made and tried out in its machine shops, the composition becoming standardized at carbon 2.5 per cent, chromium 15 per cent. Further research resulted in the Furness-Patch patent 1,206,902, for which application was made in 1915 and which issued in 1916, covering carbon 1 to 2 per cent, chromium 15 to 20 per cent.

In the course of these experiments it developed that lower carbon and higher chromium gave much better results in lathe tools than did the earlier analyses, with somewhat less distortion of pieces upon hardening. In one series of tests, roughing tools with high carbon lasted one minute, as compared with 55 minutes for similar tools with lower carbon. This was an extreme case. A more nearly normal comparison is shown in the second group below:

[1]Not verifiable.

Carbon	Chromium	Life of Tool
2.35	15.0	1 minute
0.82	17.0	55 minutes
2.32	13.15	53 seconds
1.33	19.0	4 minutes

In spite of the superior results obtained in such tools with moderate carbon, it has proved commercially desirably to keep it much higher as a precaution against the production of a softer skin through surface decarburization by unskillful hardening.

In view of their fine performance as lathe tools, it is somewhat surprising that these steels have not come into more general machine shop use. Although not equal to high speed steel in cutting speed, their average life is probably fully four times that of carbon steel. One reason for their slow spread is their much higher price; another, perhaps, is the considerably greater labor of making tools from them. Neither should be a sufficient bar to their more general use.

As Mr. Wills has given no figures for the coefficient of expansion on heating of such steels, the following data[2] may be of use:

Analysis of Material Tested:

Carbon	1.78
Manganese	.39
Silicon	.53
Chromium	18.28

Treatment	Temperature Range	Coefficient of Expansion
Annealed at 1625° F. (885° C.)	85- 400° F.	5.0 x 10 $^{-6}$
	85- 800	5.8
	85-1200	6.0
	85-1450	6.5
	400- 800	6.4
	800-1200	7.1
	1200-1450	7.6
Quenched in oil at 1850° F. (1010° C.) Drawn at 800° F. (427° C.)	18- 72	6.66
	18- 206	6.86

Although not so hard mineralogically as straight high carbon tool steels, the wear resistance of high carbon-high chromium steel is much greater. Certain forming dies of carbon tool steel, worn out after 3000 to 4000 draws, were replaced by high carbon-high chromium which remained in perfect condition after 75,000 draws. For drawing copper wire a steel of carbon 1.80 per cent, chromium 12 to 14 per cent, may be used in the annealed state. For drawing soft steel it is air-hardened from about 1560 degrees Fahr. (850 degrees Cent.). For medium hard steel it is air-hardened from about 1650 degrees Fahr. (900 degrees Cent.). In all these cases the die can be resized by cold peening and drifting. For very hard steel the plate is oil-hardened from about 1650 degrees Fahr. (900 degrees Cent.) or somewhat higher, in which case cold resizing can not be done.

[2]Midvale Research Laboratory determinations.

Their high resistance to wear reduces the machinability of these steels. The Brinell hardness should not exceed 240 or much difficulty may be met. A very little tungsten, 0.55 per cent with carbon 1.75 per cent and chromium 18.5 per cent, is enough to keep the Brinell permanently above 260. The same effect may be produced by a combination of maximum carbon and high chromium.

Considering the wide range of application of these steels, it is my belief that they will come ultimately into much more general use than at present.

Written Discussion: By A. H. d'Arcambal, Pratt and Whitney Co., Hartford, Conn.

This splendid practical paper by Mr. Wills should be of interest to many of our members due to the increased usage of these high carbon-high chromium tool steels during the past few years.

My first experience with this material was some eight years ago when our company introduced cylindrical plug gages made of 2.25 per cent carbon, 12 per cent chromium steel similar to the author's Steel #1. These gages proved to be very much superior to ordinary tool steel gages in regard to wear resistance, as well as outwearing gages made of alloy tool steels and high speed steel. At a later date we introduced chromium plated cylindrical plug gages, these chromium plated gages proving to be very much superior to the high carbon-high chromium gages in regard to wear resistance. This resulted in discontinuing the high carbon-high chromium steel cylindrical plug gages.

Our experience with two of the types covered by the author, namely, Steel #1 and Steel #3, has resulted in our adopting the Steel #3 type for certain products of our manufacture. This lower carbon, chromium-molybdenum-vanadium type is easier to machine than the higher carbon, high chromium type, hardens just as satisfactorily at a slightly higher heat and we were pleased to note from Mr. Wills' paper, is considerably tougher than the original high carbon-high chromium grade.

We pack-harden parts made of the Steel #3 type, using a temperature of 1800 degrees Fahr. (980 degrees Cent.), soaking at temperature for at least two hours. We prefer oil quenching to air cooling due not only to the slightly higher Rockwell hardness obtained in the as-quenched condition, but also the fact that air-cooled parts have a slight amount of decarburized surface, the oil-quenched specimens being file hard on the surface. We do not recommend an 1850-degree Fahr. (1010 degree Cent.) pack hardening heat for parts made of this #3 type for at times considerable austenite is developed if this temperature is very slightly exceeded, an undesirable structure in our opinion. In Table V, the author states that the structure of Steel 3 given the 1850-degree Fahr. (1010 degree Cent.) treatment is mostly austenitic. Considering that the Rockwell hardness of the as-quenched specimens is C 65, it is our opinion that only a small amount of austenite is present.

Specimens of the #3 type heated in the electric Globar furnace with atmospheric control at 1850 degrees Fahr. (1010 degrees Cent.), and properly soaked, were quite heavily decarburized whether oil-quenched or air-cooled, the surface Rockwell hardness being only C 56-57 and easily filed. Removing 0.010 inch from the surface resulted in Rockwell readings of C 64-65 on both the oil-quenched and air-cooled specimens. The test pieces used measured $1\frac{1}{8}$ inch diameter by 4 inches long.

A satisfactory fracture is obtained on Type #3 steel whether quenched from 1800 or 1850 degrees Fahr. (980-1010 degrees Cent.). This chromium-molybdenum-vanadium type of steel has a finer, more velvety fracture than the #1 type.

We would appreciate Mr. Wills' answering the following questions:

1. What method was used for hardening the various specimens?
2. Does the vanadium content in Steel #2 have a considerable effect on the impact properties?
3. Would you expect as high impact values on Steel #3 oil-quenched as were obtained on the air-cooled test pieces?
4. What were the Rockwell hardness readings on all test specimens used?
5. Would fairly large masses of the steel #3 type properly harden in air?

We hope that the author will continue with his interesting and valuable experiments with these high carbon, high chromium tool steels, publishing the data at a later date.

Oral Discussion

N. I. STOTZ:[3] I believe Mr. Wills made the statement that in working these types of steels the best results were always obtained by annealing before forging. While this is likely true under present conditions of mill operations, I would like to state that our best forging practice has always been obtained, where possible, charged from the molds at about 1500 degrees Fahr into a furnace heat at the same temperature to allow the forging temperature to slowly be raised from that point, rather than by annealing. It is not always possible to do that because of commercial conditions at the present time, but where possible I think that point is strongly in favor of the case of working this type of material on the hammer particularly.

S. EPSTEIN:[4] Mr. Wills states in the paper that it is possible to harden these steels by quenching from a high hardening temperature and get a completely austenitic structure, and then draw back, but that treatment is not advisable for this type of steel, with which I agree. I have in mind a very similar type of steel containing 2 per cent of nickel in addition to the chromium and the carbon, where you get an austenitic structure as cast. In a case like that, if you wanted to harden the steel, would it be more advisable to merely draw the "as cast" steel, or would you heat it up, quench it and then draw it back?

R. M. SANDBERG:[5] In experimenting with high carbon-high chromium steel our results were similar to those found by Mr. d'Arcambal. That is, at a higher quenching temperature an austenitic condition is produced and in subsequent tempering the volumetric changes are seriously affected, but if the proper hardening temperature is used, which is usually found to be somewhat below the temperature where this austenitic condition is formed, we can, on subsequent tempering, return the material to its original volume as found by Mr.

[3]Braeburn Alloy Steel Co., Braeburn, Pa.

[4]Battelle Memorial Institute, Columbus, Ohio.

[5]Columbia Tool Steel Co., Chicago Heights, Ill.

d'Arcambal. Most people when first starting to use this type of tool steel got the impression that it had an abnormally wide hardening range. However, anyone who has gone into the heat treatment fundamentally has quickly proven that this is not the case but that it is best to follow the fixed rules of good heat treatment with this material the same as you do with any other highly technical product of the ferrous group.

Author's Closure

Regarding Mr. Scheid's discussion, he brings up the point that it would have been desirable to have had the tests on physical properties include more than the three compositions dealt with. While we agree on this, the laboratory tests were limited to these three only as these are the main ones we have been concerned with from a manufacturing standpoint.

He also mentions that further impact values should be included with variation of hardening temperature and time. A more complete study of these steels could readily include this, for, like high speed steels, the hardening range of the high carbon-high chromium steels is wide and a comprehensive investigation should include the determination of the various physical properties after a series of hardening temperatures each followed by a similar set of drawing treatments.

Mr. Scheid refers to their finding a tungsten-silicon alloy superior to the vanadium and vanadium-molybdenum combinations as an oil or air hardening steel. We would be interested to know in what respects it has shown up better and the approximate composition.

He finally states that a lower hardening temperature on these steels would develop best physical properties. While in general the lowest hardening temperature that will develop maximum hardness should be used, our rather extensive experience has indicated a hardening range of 1650 to 1750 degrees Fahr. for the straight chromium type (#1) and 1750 to 1850 degrees Fahr. for the others (#2 and #3). We do not see how they could be much lower unless an extremely long soaking time is used. The tendency in many cases is to shorten the time or underheat.

Mr. Cox's remarks regarding the history of high carbon-high chromium steels are timely and interesting. It does seem strange that these steels are not used more for cutting tools, but in the past the high price and machinability have favored the use of high speed steel. With the lower prices now prevailing and the increased cost of high speed steel, the field of usefulness for high carbon-high chromium steels should be broadened.

Mr. Stotz mentioned a practice at their mill of taking ingots hot from the melting department (as about 1500 degrees Fahr.) and charging them into the heating furnace at about the same temperature. This is a good method from the standpoint of economy. It is more expensive to anneal the ingots before cogging, but unless the schedule is such that it can be followed out promptly it is quite difficult to do otherwise than anneal, preheat, and charge in the heating furnace.

Mr. Epstein asked a question about a high carbon-high chromium steel with 2 per cent nickel which has an austenitic structure as cast. He inquired

if it was desired to harden the steel as cast, whether it would be better to merely give the steel a draw or would you bring up to the hardening temperature, quench, and draw back.

In reply to this, it would be the best practice to give the castings a thorough anneal. They can then be brought up to the hardening temperature, quenched so as to obtain a largely martensitic structure and then given a low temperature draw.

Regarding paragraph 3 of Mr. Erb's discussion, we are pleased to note that he as a user of these steels agrees that it would be desirable if all requirements could be met with a single steel of this type, but that due to the variety of applications that have to be covered, more than one steel is required.

Answering Mr. Erb's questions in paragraph 4, the test specimens were ground sufficient to clean up the surface previous to the determination of hardness and impact values. The impact test specimens were checked for hardness although this was not given in the table. These as well as the hardness values for the compressive strength and expansion series are listed below:

Impact Values

Drawing Temp. Degrees Fahr.	Steel #1		Steel #2		Steel #3	
	Rockwell C	Ft.-Lb.	Rockwell C	Ft.-Lb.	Rockwell C	Ft.-Lb.
As quenched	66	4.6	66	8.9	63	37.0
400	63.5	8.0	61.5	20.0	61	56.0
500	61.5	5.7	61.5	21.3	58	63.8
600	59.5	5.5	59.5	21.1	59	59.2
700	57.5	6.1	58.5	19.5	58	52.4
800	57.5	7.1	57.5	22.5	56.5	46.4
900	54.5	5.0	57.5	18.7	....	....
1000	51.5	5.8	52.0	21.8	60.0	42.0
1100	47.5	9.3	46	33.3	48.0	63.0

Ultimate Compressive Strength

Drawing Temp. Degrees Fahr.	Steel #1		Steel #2		Steel #3	
	Rockwell C	Lbs. per Sq. In.	Rockwell C	Lbs. per Sq. In.	Rockwell C	Lbs. per Sq. In.
400	62.5	438,500	63.5	425,500	62.3	403,625
600	60.5	396,625	60.5	389,250	60.3	364,750
850	54	302,125	59	350,375	59.2	329,000

Expansion

Drawing Temp. Degrees Fahr.	Steel #1		Steel #2		Steel #3	
	Rockwell C	Per Lin. In.	Rockwell C	Per Lin. In.	Rockwell C	Per Lin. In.
As quenched	66.0	.00085	67.0	.00142	66.5	.00087
400	62.5	.00063	64.0	.00110	63.0	.00050
500	....	.....	62.0	.00080	62.5	.00037
600	60.0	.00061	....	.....	....	.....
750	....	.....	60.5	.00037	61.0	.00007
800	56.0	.00013	....	.....	....	.....
1000	50.0	.00030	54.0	.00042	59.0	.00050

Mr. D'Arcambal inquires in question #1 as to the method of hardening. The hardening and impact tests were run open fire (electric muffle furnace) bringing up slowly to temperature and allowing about ½ hour at temperature for the former and ten minutes at temperature for the impact tests. The compression and expansion tests were pack hardened allowing about 1 hour at temperature before quenching.

Answering his question #2, the vandium content of Steel #2 results in a material increase in impact properties over Steel #1.

As to question #3, we would not expect any great difference in impact value between Steel #3 oil-quenched and when cooled in air, although we have not checked this by actual test.

Question #4 is covered in the above table covering one of Mr. Erb's questions.

Replying to question #5, we checked the hardness on air-hardened test pieces running from about 1 pound to 25 pounds in connection with the curves shown in Fig. 1, and found that even with these larger pieces that on the edges the hardness ran Rockwell C 62 to 64, although we did not check the middle of the flat faces. For the average tool purpose it is safe to assume that fairly large masses of #3 Steel will develop satisfactory hardness when cooled in still air, this, of course, dependent on holding the proper time at the hardening temperature and avoiding excessive scaling that tends to retard the air cooling.

PROLONGED TEMPERING AT 100 DEGREES CENT. AND AGING AT ROOM TEMPERATURE OF 0.8 PER CENT CARBON STEEL

By G. A. Ellinger and R. L. Sanford

Abstract

The investigation described in this paper is a further application of the use of thermomagnetic analysis as an aid in metallographic research. Changes taking place within hardened steel during aging are of small magnitude and difficult to determine by the usual methods. Thermomagnetic analysis has been successfully applied to show the effect on hardened steel of prolonged low temperature tempering and of aging for long periods of time. Observations are given which appear to support the following conclusions:

1. Martensite is apparently unstable but tends toward stability upon the application of heat or upon aging. Liquid air cooling after quenching has an appreciable effect in accelerating stabilization.

2. Retained austenite in hardened steel is not affected by tempering but it decreases materially in amount upon aging.

3. Tempering causes no change in the rate of formation of cementite, while aging apparently greatly increases the time of its formation.

Introduction

IN a previous paper[1] there were pointed out some of the advantages and possibilities of thermomagnetic analysis as a method of studying constitutional changes occurring during the tempering of quenched steel. It was possible by this method to follow the progress of the transformations while they were occurring and some new evidence as to their nature was obtained. The method has been used in the

[1] G. A. Ellinger, "Thermomagnetic Investigation of Tempering of Quenched 0.75 Per Cent Carbon Steel," Bureau of Standards Journal of Research, Vol. 7, 1931, p. 441; RP 350; also, R. L. Sanford and G. A. Ellinger, "Thermomagnetic Phenomena in Steel," Transactions, American Society for Steel Treating, Vol. 20, 1932, p. 263.

Publication approved by The Director of the National Bureau of Standards of the United States Department of Commerce.

A paper presented before the Sixteenth Annual Convention of the Society held in New York City, October 1 to 5, 1934. Of the authors G. A. Ellinger is assistant metallurgist, and R. L. Sanford is physicist, National Bureau of Standards, Washington, D. C. Manuscript received June 29, 1934.

Table I

Treatment of Specimens After Quenching from 825 Degrees Cent. Into Water at 20 Degrees Cent.

Specimen	Treatment
1	as quenched
2	tempered 168 hours at 100 degrees Cent.
3	tempered 360 hours at 100 degrees Cent.
4	tempered 750 hours at 100 degrees Cent.
5	tempered 1500 hours at 100 degrees Cent.
6	as quenched
7	tempered 1500 hours at 100 degrees Cent.
8	tempered 1500 hours at 100 degrees Cent., and aged 2 years at room temperature (approximate range 0-35 degrees Cent.)
9	aged 2 years at room temperature
10	cooled 2½ hours in liquid air (—190 degrees Cent.)
11	cooled 2½ hours in liquid air, tempered 1500 hours at 100 degrees Cent., and aged 2 years at room temperature.

present investigation to study the aging at room temperature of quenched high carbon steel and the effect of prolonged heating at 100 degrees Cent.

Material and Methods

Cylindrical specimens 10 centimeters in length and 5 millimeters in diameter were prepared from basic open-hearth steel of the following composition:

	Per Cent
Carbon	0.78
Manganese	0.56
Phosphorus	0.020
Sulphur	0.031
Silicon	0.11
Chromium	0.06
Nickel	0.05
Copper	0.06

After normalizing, the specimens were heated in a lead bath maintained at a temperature of 825 degrees Cent. (1515 degrees Fahr.) and then quenched in water at 20 degrees Cent. In order to obtain uniform action during quenching, four specimens were heated and quenched together in a small jig, so designed as to allow free circulation of water around each specimen. After quenching, some of the specimens were immersed in liquid air for a period of 2½ hours. Details of the treatment of each specimen are given in Table I.

As an aid in conciseness and brevity specimens held 1500 hours at 100 degrees Cent. are designated as "tempered" and those kept at room temperature (an approximate extreme range of 0-35 degrees Cent.) for 2 years as "aged." In addition to the thermomagnetic

analysis which was carried out with a slightly modified form of the apparatus previously described,[2] the microstructure was examined and the hardness was determined on auxiliary specimens, notched before heat treatment so that small pieces could be broken off when desired during the course of the treatment. The hardness was determined with a Vickers hardness tester.

Observations and Results

In carrying out the thermomagnetic analysis the specimens were subjected to a constant applied magnetizing force of 277 oersteds. The dimensional ratio of the specimen was such as to bring the magnetization always into the third stage. The uncertainties usually encountered when the magnetization is in the unstable second stage, represented by the steep part of the B-H curve were thus avoided.

Two methods of procedure were used. In the first method, the specimen was heated at a uniform rate of approximately 1.5 degrees Cent. per minute and the degree of magnetization noted at suitable intervals. In the second method, the procedure consisted in heating a specimen to a definite temperature and holding it at that temperature until the magnetization reached a constant value. This was done at temperatures of 200, 235 and 300 degrees Cent., (390, 455, 570 degrees Fahr.) with several runs at 400 degrees Cent. (750 degrees Fahr.).

Fig. 1 shows the results obtained by the first method, that is, by uninterrupted heating to 300 degrees Cent. of specimens quenched and tempered at 100 degrees Cent. for various periods up to 1500 hours. The magnetization increased with temperature to a maximum which occurred at 280 degrees Cent. (535 degrees Fahr.) and then decreased with increasing temperature. The maximum was lowered progressively by tempering at 100 degrees Cent., the change being most marked in the early part of the tempering period. The abrupt increase in magnetization took place at higher and higher temperatures as the tempering period was increased.

Fig. 2 illustrates the change in hardness as tempering progressed. The part of the curves corresponding to relatively short tempering periods is indicated by dotted lines, as the trend in this region is somewhat uncertain. Several investigators have reported an in-

[2] R. L. Sanford, "Apparatus for Thermomagnetic Analysis, Bureau of Standards Journal of Research, Vol. 2, 1929, p. 659, RP 50.

crease in hardness for quenched specimens tempered for short periods of time at 100 degrees Cent. In this investigation, no hardness determinations were made on specimens tempered for less than 1 week, at which time the hardness was appreciably lower than that of the quenched material.

Fig. 3 is typical of the results of thermomagnetic analysis when the second procedure given above was followed. Here are shown the

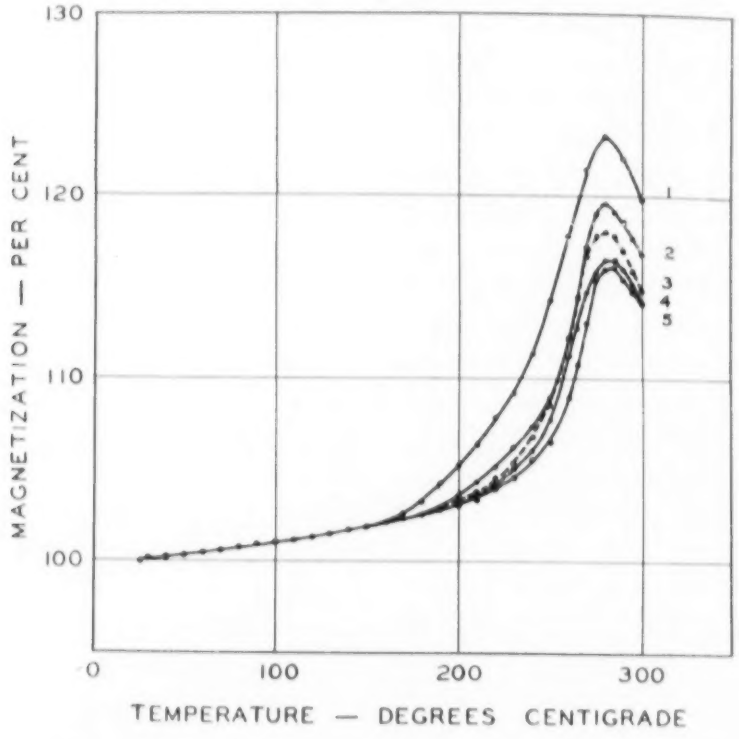


Fig. 1—Effect of Prolonged Tempering at 100 Degrees Cent. Upon Magnetization. Details of treatment of each specimen (numbers 1, 2, 3, 4 and 5) are given in Table I.

changes in magnetization which took place upon heating to, holding at, and cooling from a predetermined temperature. The magnitude of the changes observed upon holding at a constant temperature and the time required for the changes to be completed depended upon the treatment to which the various specimens had previously been subjected. All of the other curves were substantially similar in form to those shown.

In this procedure a progressive increase in magnetization took place immediately upon heating, the rate of increase depending upon the rate of heating. Upon holding at 200 degrees Cent. the mag-

netization continued to increase over a period of several hours, after which, no further increase was noted. On cooling to room temperature little, if any, change in magnetization was observed.

Upon reheating, no appreciable change in magnetization was noted until the temperature was approximately 230 degrees Cent. (445 degrees Fahr.). On holding at 235 degrees Cent. (455 degrees

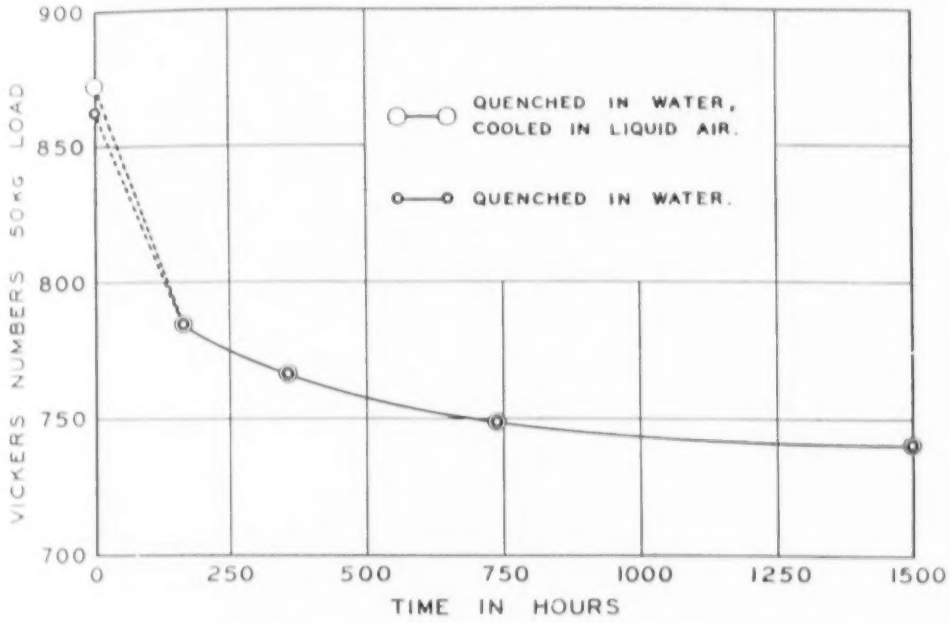


Fig. 2—Effect of Prolonged Tempering at 100 Degrees Cent. Upon Hardness.

Fahr.) an increase in magnetization was observed, which continued for approximately 40 minutes. No further change took place upon holding for an hour but in cooling to room temperature, a progressive increase in magnetization was again observed.

On reheating to 300 degrees Cent. (570 degrees Fahr.) a gradual progressive decrease in magnetization took place, which continued for some time as that temperature was maintained constant. Upon cooling, a pronounced A_0 transformation was observed which indicated the presence of cementite. Specimens reheated to 400 degrees Cent. (750 degrees Fahr.) revealed, in no case, any further change in magnetization as the specimen was held at that temperature.

From this type of curve were derived the data shown graphically in Fig. 4, in which the total changes in magnetization at each of these temperatures are plotted on a percentage basis. The value

for 200 degrees Cent. (390 degrees Fahr.) was the total change upon heating to and holding at that temperature. The values given for the other temperatures were the changes that took place only during the time the temperature was held constant.

The increases in magnetization occurring at 200 degrees Cent. (390 degrees Fahr.) show several features. The maximum increase occurred in freshly quenched specimens (No. 6) while the effect was

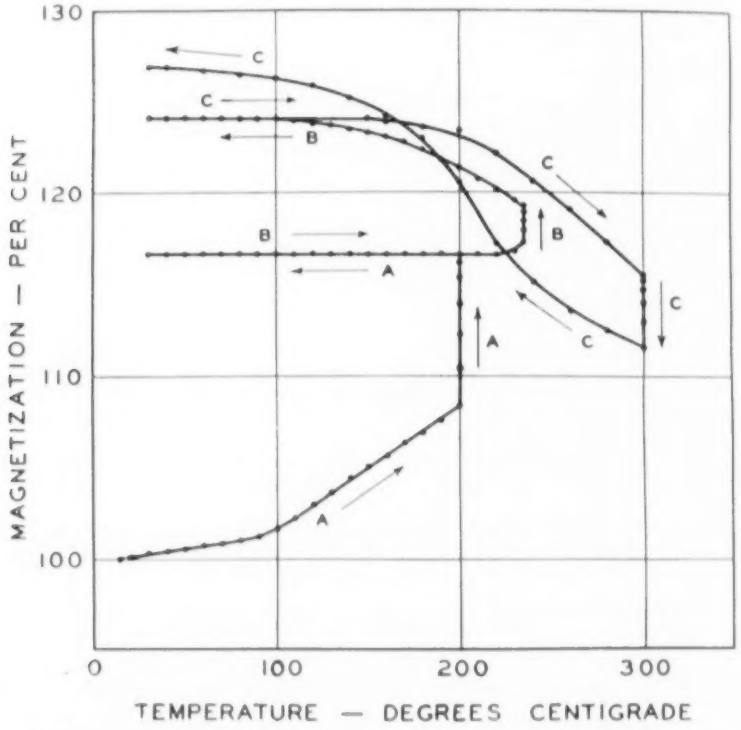


Fig. 3—Typical Thermomagnetic Curve Showing Changes in Magnetization in Quenched Steel.
A—Maximum Temperature 200 Degrees Cent.
B—Maximum Temperature 235 Degrees Cent.
C—Maximum Temperature 300 Degrees Cent.

somewhat less in specimens aged for 2 years (No. 9). Tempered specimens (No. 7) and those tempered and aged (No. 8) showed the least increase, both having practically the same value. After treatment with liquid air the increase in magnetization after tempering and aging (No. 11) was the same as that observed immediately after the liquid air treatment (No. 10).

The increase in magnetization observed upon holding the specimens at 235 degrees Cent. (455 degrees Fahr.) was affected by neither the simple liquid air treatment (No. 10) nor tempering (No.

7). Aging for 2 years (Nos. 8, 9 and 11), on the other hand, very greatly reduced the change.

The decrease in magnetization occurring when a specimen was maintained at 300 degrees Cent. (570 degrees Fahr.) was apparently

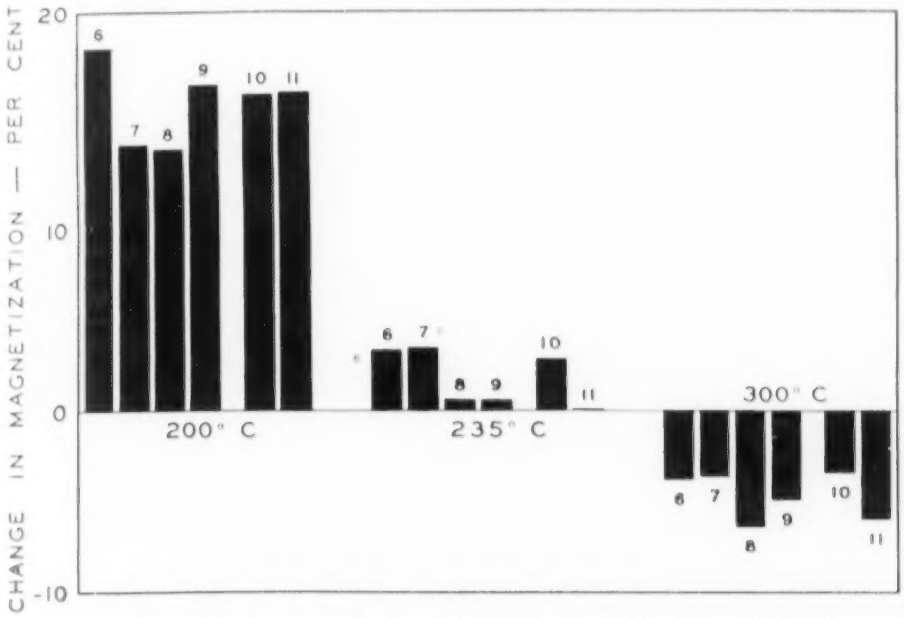


Fig. 4—Changes in Magnetization Taking Place at 200, 235 and 300 Degrees Cent. Details of treatment of each specimen are given in Table I.

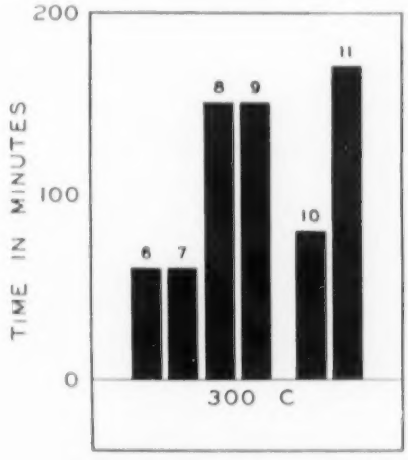


Fig. 5—Time Required for Decrease in Magnetization to be Completed at 300 Degrees Cent.

affected very little by tempering (No. 7) but aging (Nos. 8, 9 and 11) influenced this change materially. The maximum change was noted in the aged specimens regardless of previous treatment. Fig. 5 shows

the differences in time required for the changes occurring at 300 degrees Cent. (570 degrees Fahr.) to be completed. Specimens aged 2 years at room temperature (Nos. 8 and 9) required more than twice as long for completion of the change as did the freshly quenched specimens (No. 6). In both cases the liquid air treatment (Nos. 10 and 11) increased the time required.

A study of the microstructure of the quenched and tempered specimens was made. Typical structures are shown in Figs. 6 to 10, inclusive. The martensite needles showed a progressive darkening as time of tempering increased and a gradual breaking down, particularly in the latter stages, until after 1500 hours it was difficult to reveal structural details with any degree of sharpness.

Discussion

From the results it is evident that, even at room temperature, quenched steel undergoes progressive changes, the magnitude of which is determined by preliminary treatment such as cooling in liquid air or heating for prolonged periods at 100 degrees Cent. It is apparent further, that the effect of these changes differs from the "age hardening" phenomena in other metals, but rather is a softening effect.

In considering the changes at 200 degrees Cent. (390 degrees Fahr.), it was observed that the maximum increase in magnetization occurred in freshly quenched specimens and was least in tempered specimens. These changes have previously[3] been attributed to the gradual release of stresses induced by quenching.

It is known from the results of other investigators that martensite is in a highly stressed state, the stresses being so high as to convert the normal cubic lattice of alpha iron into a pseudo-tetragonal form. Furthermore, martensite is in the state of a supersaturated solid solution, such solution being forced by the extremely rapid cooling with no chance for the precipitation of carbon and the formation of cementite.

Low temperature tempering or sufficiently long aging periods are believed to relieve these stresses materially through the gradual precipitation of carbon, either as atomic carbon or some constituent other than cementite. The lattice then assumes more nearly its cubic dimensions with a resultant decrease in hardness. Likewise, the stability of the tempered or aged martensite is increased.

[3] G. A. Ellinger, loc. cit.

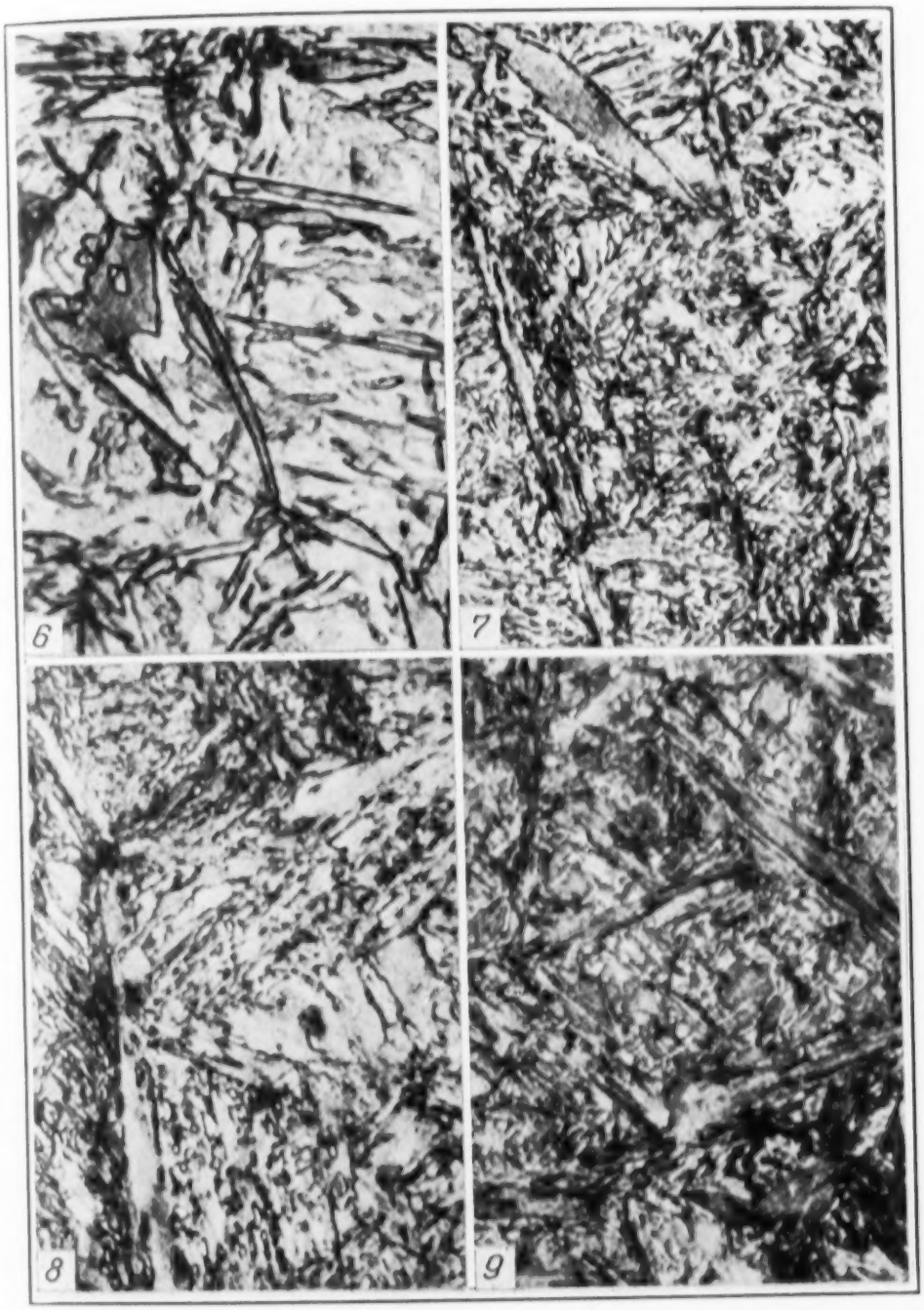


Fig. 6—Structure of Specimen Freshly Quenched in Water, × 1500.
Fig. 7—Structure of Specimen After Tempering 168 Hours at 100 Degrees Cent., × 1500.
Fig. 8—Structure of Specimen After Tempering 360 Hours at 100 Degrees Cent., × 1500.
Fig. 9—Structure of Specimen After Tempering 750 Hours at 100 Degrees Cent., × 1500.

The effect of cooling specimens in liquid air is quite pronounced. The magnitude of the change at 200 degrees Cent. was affected by neither tempering nor aging, which indicates that little significant change took place in the martensite. Greater stresses must have been induced by cooling in liquid air than in water quenching and certain of these stresses must have been released upon

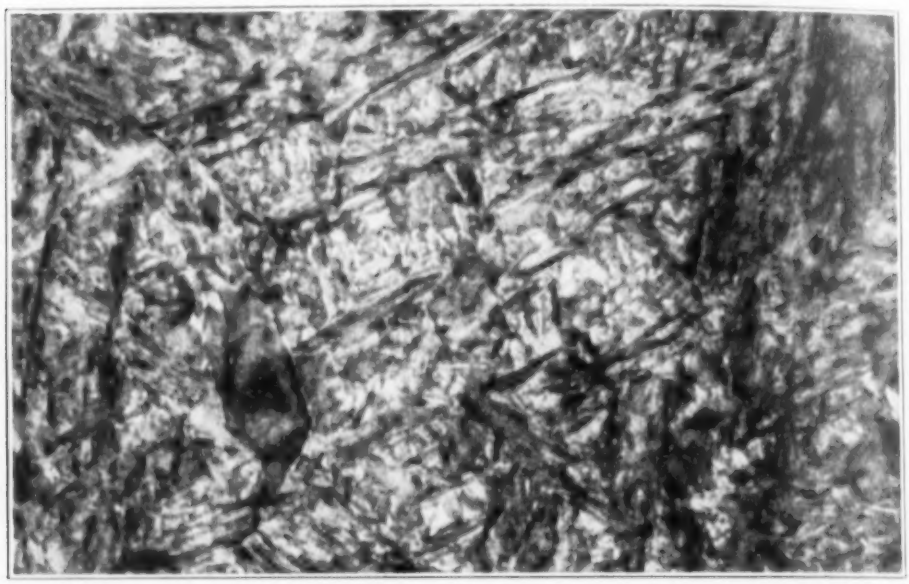


Fig. 10—Structure of Specimen After Tempering 1500 Hours at 100 Degrees Cent., × 1500.

subsequently warming to room temperature. This release must have been of considerable magnitude as evidenced by the diminished increase of magnetization at 200 degrees Cent. (390 degrees Fahr.). The effect of liquid air treatment, not unlike that of tempering, appears to be one of stabilization of martensite. This is further indicated by the fact that tempering and aging apparently have no effect upon the magnitude of the magnetic change at this temperature.

Small amounts of austenite usually exist in freshly quenched steel. This constituent is unstable, being forcibly detained in the undercooled state and will transform to stable phases as soon as permitted. That such was the case is supported by the data. The freshly quenched steel evidently contained an appreciable amount of austenite which decomposed upon heating to 235 degrees Cent. (455 degrees Fahr.). Even after tempering there was little or no

change in the amount of austenite present, but after aging for 2 years this amount was reduced materially. Those specimens which, in the freshly quenched condition, showed the presence of austenite increased somewhat in magnetization upon cooling from 235 degrees Cent. The nature of the curves, Fig. 3, suggests that a small amount of cementite was formed at this point, most of which probably came from the decomposition of austenite.

The decrease in magnetization taking place at 300 degrees Cent. (570 degrees Fahr.) is attributed to the formation of cementite by the combination of carbon precipitated from martensite with alpha iron. As this took place above the A_o transformation temperature of cementite (normally considered as 215 degrees Cent.), that portion of alpha iron which combines with carbon is rendered ineffective magnetically with a resultant net reduction in the total magnetization. A hypothesis was set up previously,[4] indicating that in the range from 235 to 300 degrees Cent. (570 degrees Fahr.) atoms of carbon precipitated from the supersaturated martensite solution would unite with atoms of iron to form cementite.

Tempering or cooling in liquid air apparently had little effect upon the magnitude of the changes taking place at 300 degrees Cent. On the other hand, aging for a considerable period of time produced a noticeable effect. The magnitude of the change and its duration were materially increased.

The evidence points to a more rapid formation of cementite from the decomposition products of freshly quenched martensite than from those of aged martensite. The amount of carbide formed depends upon the amount of carbon in the steel and should be the same for a given percentage of carbon regardless of its form. A certain amount was formed while heating to 300 degrees Cent. and the remainder while holding at that temperature. The intensity and duration of the change at 300 degrees Cent. was greater in aged than in freshly quenched specimens. This leads to the belief that in aged specimens less carbide was formed in the range between 235 and 300 degrees Cent. than while holding at 300 degrees Cent., while in unaged specimens most of the carbide is formed between 235 and 300 degrees Cent. The sluggishness indicates the possibility that carbon atoms are in less favorable position to unite with iron atoms after aging and require considerably more time for the reaction to take place.

[4] G. A. Ellinger, loc. cit.

Summary

Thermomagnetic analysis has been applied to the study of 0.8 per cent carbon steel quenched in water and subsequently tempered 1500 hours at 100 degrees Cent. and aged at room temperature (0 to 35 degrees Cent.) over longer periods of time. Some quenched specimens were cooled in liquid air for 2½ hours and subsequently tempered and aged. Changes took place during heating of these specimens as follows:

1. An increase in magnetization occurred during the heating of the material and while it was held at 200 degrees Cent. (390 degrees Fahr.) the magnitude of which in quenched specimens, was appreciably affected by previous tempering, aging, and cooling in liquid air. Tempering and aging had no effect on specimens previously cooled in liquid air.

2. An increase in magnetization occurred in specimens held at 235 degrees Cent. (455 degrees Fahr.), which was not appreciably changed by previous tempering or liquid air treatment but was materially changed by aging.

3. A decrease in magnetization was observed in specimens held at 300 degrees Cent. (570 degrees Fahr.), which was changed appreciably in both magnitude and duration by previous aging.

The explanations offered for the behavior of this material after tempering and aging, based upon observations of changes taking place at various temperatures, are as follows:

The carbon of martensite is gradually precipitated both after prolonged tempering at 100 degrees Cent. and after aging, so that the more stable cubic form results. Martensite is also changed by cooling in liquid air but is not further changed by either tempering or aging.

The small amount of austenite retained in 0.8 per cent carbon steel after quenching is not affected by tempering but is decreased materially by aging. Liquid air treatment has no apparent effect upon this retained austenite.

Between 235 and 300 degrees Cent. (455-570 degrees Fahr.), in what is believed to be the cementite formation range, previous tempering causes no observable difference in magnitude or time. Aging, on the other hand, increases both magnitude and time of the reaction.

ACKNOWLEDGMENT

The authors wish to express their appreciation and gratitude for the conscientious assistance and co-operation of E. G. Bennett of the National Bureau of Standards, in making many of the determinations in this investigation.

DISCUSSION

Written Discussion: By R. H. Harrington, General Electric Research Laboratories, Schenectady, N. Y.

Thermomagnetic analysis is a relatively uncommon tool of research in metallurgy. Therefore this report of a research employing this tool is of great interest and value to the science of metallurgy.

There is still a wide diversity of opinion as to the condition of the carbon in solution in austenite and in freshly quenched martensite. The results of this research indicate a strengthening of the belief that carbon exists in atomic solution in the gamma iron lattice of austenite, and that the tetragonal form of martensite may involve an atomic precipitation of carbon.

Recognizing the definitions of "tempering" and "aging," as used by the authors, the writer believes there might be some qualifying of these statements in the summary:

1. "The small amount of austenite retained in 0.8 per cent carbon steel after quenching is not affected by tempering, but is decreased materially by aging."

2. "Between 235 and 300 degrees Cent. in what is believed to be the cementite formation range, previous tempering causes no observable difference in magnitude or time."

From various sources in the literature it would appear that the rate of transformation of retained austenite is exceedingly slow at 100 degrees Cent. and quite appreciable at 180 degrees Cent. The rates of transformation at 100 degrees Cent. and at room temperature may be nearly identical. The authors compare the results of tests on samples after 63 days at 100 degrees Cent. and after 730 days at room temperature. It might, therefore, be interesting to compare tests of samples tempered 63 days at 100 degrees Cent. with those of samples aged 63 days at room temperature.

Oral Discussion

DR. C. NUSBAUM:[1] I am pleased to have this opportunity of making some comments on this method of analysis, especially in view of the fact that I think it is one of the most useful methods of research which has been almost entirely neglected by metallurgists. If you look through the literature you will find that Honda has done most of the work in thermomagnetic analysis. Some work has been done in England, but it was not until 1920, fourteen years ago,

[1]Case School of Applied Science, Cleveland.

that investigations along this line were undertaken in this country. The apparatus which was used in the present experiment was designed and constructed at the Bureau of Standards in the year 1920 and has since been used in several investigations, of which this is perhaps the most extensive.

I would like to ask a question in regard to the magnetic measurements that were made preliminary to the analysis itself. It is assumed that all the samples were initially measured under what is called the same magnetizing force and that they all within limits, had the same magnetic induction; in other words, that if one measured the absolute value, all the initial points for all the samples agreed and coincided with each other. If this is not the case, then unfortunately the given per cent of increase does not mean very much, and every individual curve would have to be given. If the initial values do not agree, this would displace all these points relative to each other and the final values might agree with each other much more closely or much less. Without a statement as to the initial condition of each sample the data are definitely open to criticism.

I agree in regard to the general interpretation of the transformations of Fig. 3. One may call them transformation A, transformation B, and transformation C. I do feel, however, that the curves would have been even more significant if an additional curve had been taken in which this operation was continuous. I believe it would have given us a truer measure of the magnitude of each of the transformations.

I would like to ask as to whether any X-ray study has been made or is being made by any of the men at the Bureau on this problem. It would seem that this problem, along with several others, offers an unusual opportunity for a study by means of X-rays. These changes should be very definitely indicated in any suitable X-ray picture, and probably additional information as to the nature of the transformations be obtained.

In concluding, I would like to say that many feel that microscopic and thermal analyses tell the whole story in metallography and metallurgy. Personally I have been convinced for a long time that these methods are only two of the useful tools and that every complete metallurgical laboratory should contain in addition apparatus for thermomagnetic and X-ray analysis.

E. H. ENGEL:[3] With reference to the magnetization changes shown in Fig. 3, I should like to ask the authors whether they have used X-ray analysis or any other methods to differentiate definitely between stress relief and austenite decomposition at the temperatures 200 and 235 degrees Cent. When held long enough at 200 degrees Cent., some, if not all, of the retained austenite (probably 10-15 per cent) is likely to decompose and the change in magnetization at 200 degrees Cent. may not be entirely due to the relief of stresses, just as the increase in magnetization at 235 degrees Cent. may not be entirely due to the decomposition of retained austenite.

With regard to the effect of the liquid air treatment, I agree with the authors that this statement must induce stresses in the martensite much higher than those existing after the water quench. I wonder if the authors have recorded the change in magnetization of the specimen, not only from room temperature up to 200 degrees Cent., but also from temperatures near that of

[3]Research laboratory, United States Steel Corporation, Kearny, N. J.

liquid air to room temperature. If the authors had included the total change in magnetization of a specimen as heated from —180 to 200 degrees Cent., the effect of the liquid air treatment might have been demonstrated more clearly.

DR. H. STYRI:[3] We might throw some side lights on these investigations by another method that we have employed in the SKF laboratory. We have used a dilatometer and have, since I obtained a copy of this paper, treated some bars of the same type steel at 100 degrees Cent. and at 200 degrees Cent. for a long time and measured the change in length continuously. At 100 degrees Cent. we get a very slow contraction to a minimum in about four weeks and it takes much longer time before it starts to increase again. At 200 degrees Cent. the drop is very much faster and there is a decided increase after passing the minimum.

We have made similar tests on other specimens, mostly at 200 degrees Cent. and over, and with the kind help of the research laboratories of the U. S. Steel Corp., it was determined by X-ray analysis that austenite which was present in the beginning at 200 degrees Cent. gradually disappeared. There must be some connection between the length changes and the changes in crystal structure. I am sorry that Mr. Sanford has not measured the magnetic properties at constant temperatures and recorded them against time. It would probably have given some more information.

Authors' Closure

The authors are appreciative of the interest taken by those who have offered discussion.

In most of the discussions, the use of other methods such as X-rays, the dilatometer and the like has been suggested for the purpose of checking the conclusion, generally questioned, that there is no decomposition of austenite at 200 degrees Cent. There is no doubt that useful data would have been obtained if the application of these methods had been feasible. On the other hand, the conclusions reached by various investigators using these methods are not in perfect agreement. It was our thought to add such evidence as might be obtained by thermo-magnetic analysis to that already available.

The belief that austenite was not decomposed at 200 degrees Cent. in our experiments is based upon the following considerations: After the change at 200 degrees Cent. had been completed, no further change was observed as the temperature was raised until about 235 degrees Cent. was reached. If, as has been suggested, the decomposition of austenite had been started at 200 degrees Cent. it seems reasonable to suppose that it would be completed at that temperature if given sufficient time.

Furthermore, since no progressive change was observed in the interval from 200 to 235 degrees Cent. even with the slow heating rate of approximately ¼ degree Cent. per minute, the conclusion seems obvious that whatever change did take place at 200 degrees Cent. was completed at that temperature. Upon cooling from 200 degrees Cent. no trace of the A_0 transformation was observed and consequently no appreciable amount of carbide could have been formed. It is well known that freshly quenched martensite is in a condition of

[3]SKF Laboratories, Philadelphia.

strain and consequently the magnetic change was attributed to the relief of strain.

Upon holding at 235 degrees Cent. the magnetization increased in value. The formation of carbide leads to a decrease in magnetization. The only reasonable explanation of the increase, therefore, is the decomposition of austenite. No doubt some carbide was also formed because, upon cooling, there was some evidence of an A_o transformation. The changes at 200 and 235 degrees Cent. are distinct from each other, and no progressive change in magnetization was noted between these two temperatures. The increase in magnetization at 235 degrees Cent. can be attributed to the decomposition of austenite. It seems reasonable to believe therefore that no decomposition of austenite took place at 200 degrees Cent. This is in accordance with the views of some investigators who set the temperature at which decomposition begins as high as 260 degrees Cent. If these conclusions are accepted, the others would appear to follow as a logical consequence.

The representation of the results in terms of percentages is a matter of convenience and seems justified, since no conclusions are based on absolute values of induction or permeability. The initial values for all specimens similarly treated were approximately the same. This was indicated by the original magnetometer readings.

After some experiments in which runs were made without cooling to room temperature between successive holding temperatures, it was decided that cooling between holding temperatures is preferable. This procedure, which is more convenient, also provides a means for demonstrating whether or not carbide is formed at a given temperature.

NEW DEVELOPMENT IN ELECTRICAL STRIP STEELS CHARACTERIZED BY FINE GRAIN STRUCTURE APPROACHING THE PROPERTIES OF A SINGLE CRYSTAL

By N. P. Goss

Abstract

It has been found by experiment that single crystals of silicon steel or pure iron have extremely low hysteresis losses in certain crystallographic directions; however, such single crystal structures have the disadvantage of having exceedingly high eddy current losses when used in alternating current apparatus, as transformers, for example.

A new electrical strip steel has been developed which is fine-grained and has single crystal properties. These properties are obtained by the proper distribution of hot rolling, heat treating and cold rolling. The whole idea is based on a radical departure from present day theories regarding the ferromagnetic properties of single crystals and fine grain aggregates.

The final strip is characterized by a grain structure in which the crystals are quite perfect in structure and are distributed entirely at random and the magnetization curves of these fine-grained aggregates have the same symmetry and magnitude as the single crystal of the same substance.

This paper will show in some detail the principles which must be adhered to in order to develop single crystal properties in fine-grained ferromagnetic materials.

Introduction

THE greatest improvements in the magnetic and electrical properties of silicon steels have been made by refinements in the melting practice in the open-hearth furnace and the subsequent hot working and final heat treatment. These refinements have reduced the watt loss to a rather low value in many instances. This has always been accomplished by increasing the silicon content to a relatively high value; above 4 per cent. Watt losses as low as 0.32 watts per pound

A paper presented before the Sixteenth Annual Convention of the Society held in New York City, October 1 to 5, 1934. The author, Norman P. Goss, is a member of the society and is physicist in charge of metals research for the Cold Metal Process Co., Youngstown, Ohio. Manuscript received July 5, 1934.

have been obtained in the 6 per cent silicon steels, but the maximum permeability and the saturation of the iron are markedly reduced and from a mechanical standpoint these steels are extremely brittle. This practically precludes their use in electrical machinery.

As far back as 1900 Barrett, of Brown and Hatfield,[1] believed that the best magnetic steels should not have a silicon content of over 3 per cent because this quantity of silicon increased the magnetic softness to such an extent that the coercive force and retentivity were reduced to nearly half of that of commercial iron. They also found that the maximum permeability was higher in these steels than in the higher silicon steels. When the silicon content is greater than 3.5 per cent it becomes difficult to cold roll on a commercial basis. With the continued improvement in the physical properties of the hot-rolled strip material it will be as easy to cold roll the higher alloys (4 per cent silicon or more) as it is to cold roll the 3.5 per cent alloy at the present time.

When these steels are cold-rolled on the single stand 4-high roller neck bearing mill they roll as readily as any low carbon steel and in some instances a 3.5 per cent silicon steel has been rolled at a speed of 1100 feet per minute. After severe cold rolling these silicon steels are as ductile as severely cold-rolled low carbon steel.

The present investigation was directed mainly to a study of silicon steels containing up to and including 3.5 per cent silicon, and most of the work was performed on the 3 to 3.5 per cent silicon strip but the method developed can be applied to the higher silicon steels.

The idea of the cold rolling of silicon steel is not new. Many attempts have been made to do this over the last twenty years and in most cases the results obtained in the final product were not encouraging because the electrical and magnetic properties were found to be inferior to the silicon steels of the same content made by the regular hot sheet method, and also in view of the fact that the best results would naturally be expected in the 4 per cent or higher range. These steels, due to their extreme brittleness, are difficult to roll on a commercial basis.

When the 3 to 4 per cent silicon steels were cold-rolled on two-high mills it was always necessary to give many intermediate anneals before the final gage could be attained. This made the material very expensive to manufacture. The best results obtained, at least ac-

[1]Greiner, Marsh and Stoughton, "The Alloys of Iron and Steel," McGraw-Hill Publishing Co., New York City, p. 332.

cording to the published literature, is the work that has been done by Pomp.[2] He succeeded in cold rolling narrow high silicon steel strip containing over 4 per cent silicon and when properly heat treated obtained a watt loss of about 0.54 watts per pound. He found that the best results were obtained when the strip was given repeated cold rollings followed by annealing. This, he believed, caused an increased conversion of carbide to graphite with subsequent improvement in the magnetic properties. His results are contrary to those obtained by several leading producers of this material.

Present Investigation

The object of this paper is to show that by a suitable combination of heat treatment and cold rolling a greatly improved silicon steel can be obtained which has magnetic and electrical properties far in excess of the present day electrical sheet of the same silicon content or even the much higher silicon steel made by the regular hot roll method.

The whole idea is based on a radical departure from present day theories regarding the ferro-magnetic properties of single crystals and fine-grained aggregates. When hot-rolled silicon steel strip is heat treated and cold-rolled under certain controlled conditions, a fine-grained silicon steel strip, having a magnetic moment approaching that of a single crystal can be made. What is meant by magnetic moment of a single crystal will now be illustrated.

It has long been known that single crystals of ferro-magnetic materials, when studied by various methods, are characterized by a high magnetic moment. In a small or fine-grained material of the same substance, it has hitherto been assumed to have a zero magnetic moment; and if a small magnetic moment was found it was always considered to be due to the residual preferred orientation. In this field of research the work of P. Weis, Honda and Kunz[3] are the most outstanding. They have shown by various methods that the magnetization of single crystals varies depending upon the direction in which it is magnetized. In K. Honda's book, "The Magnetic Properties of Matter," a complete description of one method of investigation is given on page 154. This work was done by S. Kaya, Y. Mashiyama and Honda. They prepared a number of single

[2] A. Pomp, "The Cold-Rolling of High Quality Steel," *Stahl und Eisen*, Vol. 44, 1924, p. 1694-1696.

[3] "National Research Council Bulletin Theories of Magnetism," Vol. 18, p. 165.

iron crystals in the form of oblate ellipsoids in which the major and minor axis coincided with various important crystallographic directions. The flat plane of these ellipsoids coincided with such important crystallographic planes as the (1 0 0), (1 1 0) and (1 1 1). The magnetization of the various crystals was measured by the ballistic method and the results are shown in Fig. 1. These experi-

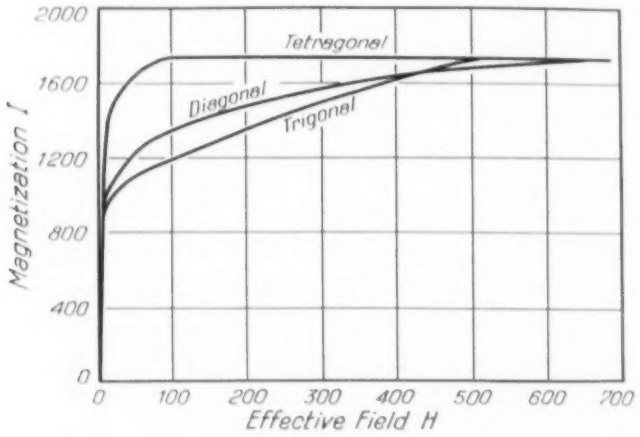


Fig. 1—Directional Magnetization in Iron Crystals.

ments proved conclusively that the intensity of magnetization depends on the direction in which a single crystal is magnetized. Their results proved that these crystals were most easily magnetized in the [1 0 0] direction or cube edge. These results have been substantiated by Ruder, Yensen and other leading authorities on the subject. Ruder in particular presented an article in TRANSACTIONS of the American Society of Steel Treating in July, 1925, which substantiated and extended the results and findings of K. Honda and his associates, though Ruder's method of investigation differed from theirs.

Similar results were obtained by Karl Beck[4] in Professor P. Weis's laboratory, in which single iron crystals made by the Goldschmidt process were investigated. He used a method differing from either that of Honda or Ruder. This method, devised by P. Weis, in which single crystals in the form of circular disks were studied is known as a torsion magnet. The results of his investigation on various kinds of ferro-magnetic single crystals agreed with that of the other investigators. The writer has applied this particular method of testing ferro-magnetic single crystals to the study of fine-grained

[4]K. Beck, Dissertation, Zurich, 1918.

silicon steels with great success. This important method may be described briefly as follows:

The single crystal or fine-grained aggregate in the form of a thin disk, about the size of a dime, is suspended in a magnetic field by means of a torsion suspension. The magnetic field is rotated about the disk which is parallel to the floor and at regular intervals the deflections of the couple are observed by means of a telescope and scale. The suspension is calibrated by means of a single crystal of silicon steel and the deflections are plotted against the various angle readings. In Fig. 2 is a photograph of the torsion magnet used.

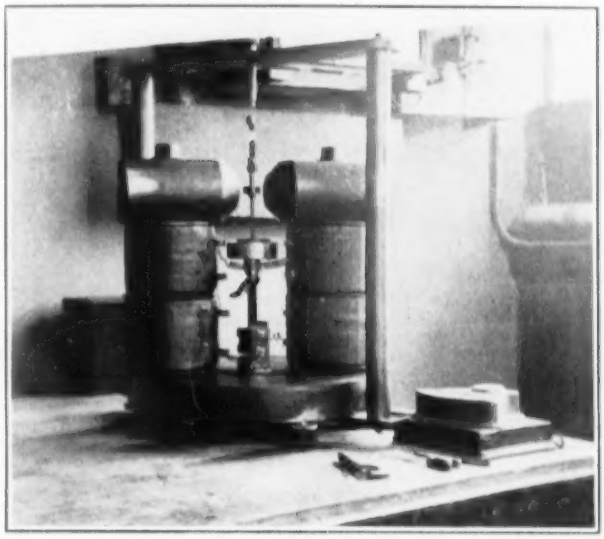


Fig. 2—Torsion Magnet Used in these Studies.

Fig. 3 is a torsion magnetization curve characteristic of a single crystal of silicon steel about 3 per cent silicon content. This particular crystal was removed from a regular 0.014 inch silicon steel strip and was about $\frac{3}{32}$ inch in diameter. It is characterized by high magnetic moment and a curve of definite symmetry which is similar to curves shown in Honda's book, page 156, and in the Bulletin of the National Research Council, Vol. 18, p. 188, by Professor J. Kunz. The symmetry of the curves of single crystals are always related to the orientation of crystals being studied. In nearly all of the silicon steel single crystals investigated in our laboratory, curves having the form of symmetry shown in Fig. 3 were obtained. These silicon steel single crystals were grown in silicon steel sheets and carefully removed. When the magnetizing field H coincides

with the path of easiest induction, no deflection will be observed; for as the magnet is gradually rotated past the point of zero deflection the crystal has a tendency to follow the field of the magnet and is only prevented in doing so by the restoring force of the couple. The maximum deflection is, therefore, only limited by the stiffness of the suspension fibre. The maximum deflections on either side of the zero positions are noted and the magnitudes of these deflections are proportional to the magnetic moment of the crystal.

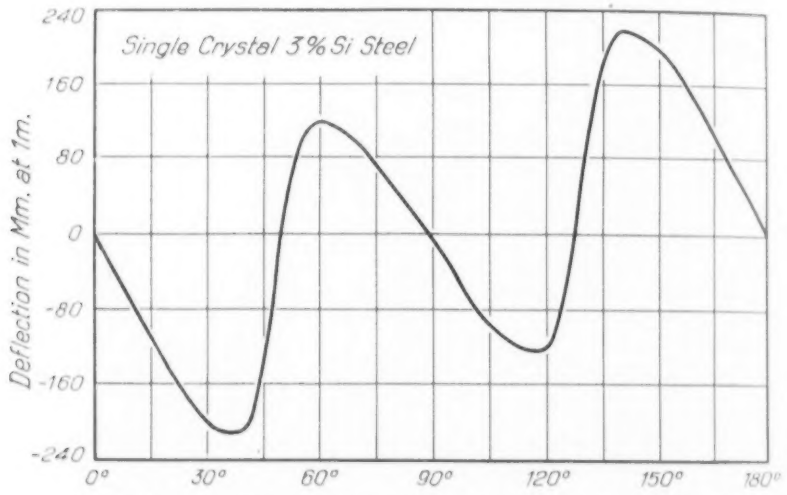


Fig. 3—Torsion Magnetization Curve of a Single Crystal of 3 Per Cent Silicon Steel.

The relationship between the magnetic moment, intensity of magnetization and permeability are clearly described in Honda's book,[5] pages 5 and 6. A mathematical relationship does exist between these quantities and these can be demonstrated physically by the torsion magnet and other magnetic testing equipment, but the torsion magnet is fundamental. The greater the deflection for any given specimen, be it single crystal or a fine-grained aggregate, the higher the intensity of magnetization; and the greater the permeability.

All the work carried out in our laboratory was done using the torsion magnet as a basis of investigation. The suspension after calibration was never changed, neither was the field strength varied during the entire duration of the work. The field strength used was entirely arbitrary; a specimen was suspended in the field of the magnet and the field strength increased until no further deflection

[5]Loc. Cit.

of the specimen could be observed with increasing field strength. In other words, the field was operated at saturation.

In Fig. 3 the torsion magnetization curve of the single crystal is characterized by four maxima in a rotation of 180 degrees of the magnet. Two of these peaks are about twice as high as the other two. This means that the crystal possesses two important directions in which magnetization is most intense; these directions are at right angles to each other,—the one having about twice the permeability of the other. As has already been stated above, this form of magnetization curve is typical of nearly all single silicon steel crystals which are grown in silicon steel strip or sheet. Two other types of torsion magnetization curves have been found especially in the higher silicon steels; however, they have proven to be exceptions. In fine-grained silicon steels made by the regular hot-rolled method the torsion magnetization curves have the same symmetry but the magnetic moments are of small magnitude.

So far it has been shown that a single silicon steel crystal is characterized by a high magnetic moment and the symmetry of the curve depends primarily on the orientation of that crystal with respect of the surface of the sheet.

Up until the present time it has been assumed that when a disk of fine-grained silicon steel the same size as the single silicon steel crystal is suspended in a torsion magnet and the field of the magnet rotated about the suspension it should experience a zero couple or twist. In other words, such a specimen would be characterized by the possibility of being homogeneously magnetized in all directions but permeability tests on hot-rolled sheet always have a greater permeability in the direction of rolling. On the other hand, it was also assumed that if a small couple did exist in such an aggregate it would mean that the small grains or crystals were not entirely oriented at random and directional characteristics existed.

A large number of such fine-grained specimens were examined by this method. The hot-rolled sheets were obtained from all of the leading manufacturers of 14 mil-material used in high grade transformers and in not one instance was a zero torque found. Specimens cut from the same sheet very often were found to have a relatively large variation of maximum deflection but always very small when compared to a single crystal of the same material. The question which remained opened to controversy, was, did all of these

specimens which were tested in a torsion magnet have directional properties?

X-Ray Investigation of Torsion Magnet Specimens

To prove whether or not preferred orientation existed in these specimens after testing in the torsion magnet all of them were subjected to X-ray examination, to determine whether or not preferred orientation of the crystals existed in these torsion magnet specimens. In view of the fact that none of them exhibited a zero torque

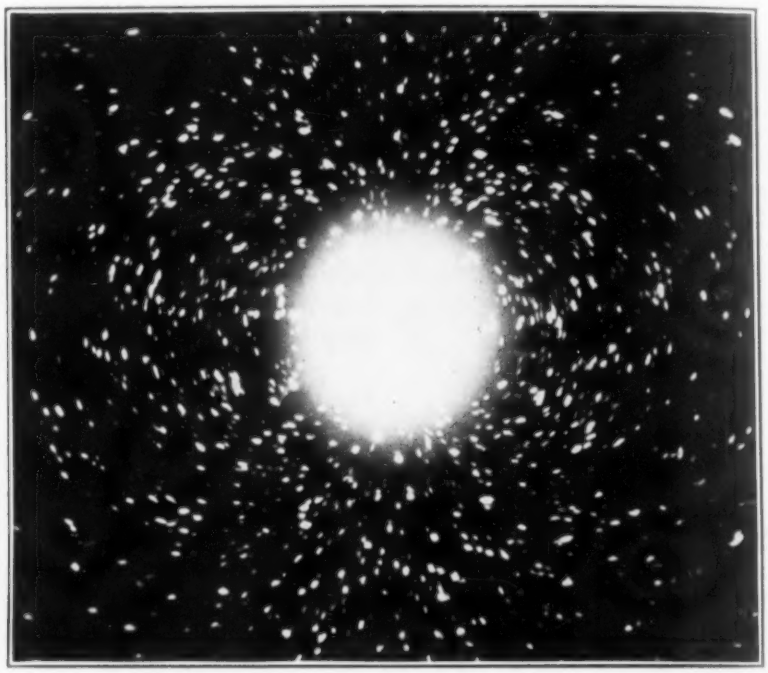


Fig. 4—X-Ray Laue Diagram of Small-Grained Silicon Steel Made of Regular Hot-Rolled Sheet. The Grains are Oriented at Random, but a Definite Magnetic Moment Exists. It is of Small Magnitude.

in the torsion magnet but differed only in magnitude, one would expect that in case preferred orientation did exist, then it should vary in the same ratio as the torsion magnetization curves. In other words, from a crystallographic point of view one should expect to find a varying degree of preferred orientation in the various specimens. The X-ray beam was transmitted at right angles to the surface of the specimens and the specimens were carefully prepared by etching them to 0.005 inch thick in a dilute solution of nitric acid; this was done to prevent strain during preparation. To make the X-ray investigation as conclusive as possible, the specimens were separated

into two groups. One group contained only those specimens which showed a relatively large magnetic moment and the other group a relatively smaller magnetic moment. All of these specimens were in the annealed condition. In the specimens which exhibited a relatively large magnetic moment not the slightest trace of preferred orientation could be detected. The X-ray definitely proved that a magnetic moment could exist even in a small-grained silicon steel characterized by a crystal structure having a random orientation. Fig. 4 is the X-ray diagram of one of these specimens and Fig. 5 is its torsion magnetization curve, yet the X-ray diagram conclusively shows that the crystal structure is random. The specimen in the case of Fig. 4 was moved at regular intervals during exposure in order to bring as many grains into the reflecting position as possible. If the X-ray had not contributed another thing to the problem it proved its great value in supplying this necessary information which formed the basis of the entire investigation. With this important discovery as a basis it was believed that it should be possible to develop in a fine-grained material magnetic moments of a greater magnitude than found in the present day small or fine-grained silicon steel hot-rolled sheet. However, at the beginning of this investigation it was hardly feasible to assume that the magnetic moment could be made to approach that of a single crystal, but this has been accomplished and can be produced on a commercial basis.

The fundamental results obtained with the aid of the X-ray spectrograph and the torsion magnet have made it possible to produce a small or fine-grained silicon steel having the characteristics of a single crystal and even in some cases exceeding the characteristics of a single crystal in some respects. Not only have these two methods of research been able to aid in the development of this new electrical strip steel but they have turned out to be indispensable as instruments of control in production.

Method of Processing Electrical Strip Steel Having Single Crystal Characteristics

Before any attempts are made to process the hot-rolled strip in production it is known ahead of time just what one is to expect as to the watt loss and permeability of the various coils of hot-rolled strip.

A small quantity of the hot-rolled strip is studied in the labo-

ratory, using several of the most probable methods which will impart to the steel the maximum properties and the procedure which gives the best results is then applied to the whole coil in the mill.

The basic principles for the production of fine-grained silicon steel having the magnetic and electrical properties approaching that of a single crystal are at this time firmly established and can be applied to the low and high silicon steels or similar magnetic alloys.

When these principles are followed it will be found that the latitudes in cold rolling and heat treating are not critical but the general principles must be rigidly followed in order to obtain the optimum results; that is, if in the laboratory we should find the best heat treating temperature at 1700 degrees Fahr., then the variation in the mill heat treating furnace can be anywhere from 1680 to 1740 degrees Fahr. and results will be just as good. The cold rolling has about the same degree of variation. If we would find that 0.030 inch is the best intermediate cold rolling gage, then ± 0.002 inch will not affect the results in the final material; however, the general practice is to adhere to the laboratory conditions as closely as possible.

It might be pointed out that the properties of this new material are not influenced by the method of cold rolling. The same results can be obtained whether the rolling is on a regular two-high or four-high mill, however, from a commercial point of view the Steckel type of four-high cold rolling mill with small work rolls has the advantage in that it is able to cold roll at greater rolling speeds with greater reductions than the regular two-high mill and the chances of failure during cold rolling are reduced to a remarkably small amount.

In view of the experience of others it was decided not to cold roll silicon steel with a silicon content greater than 3.5 per cent, again there were many advantages to be gained if a good electrical strip could be developed in the 3 to 3.5 per cent silicon range. These steels of lower silicon content cost less to make in the open-hearth than the 4 per cent. Their magnetic saturation is higher and their maximum permeability and the permeability at higher inductions are greater. The material is much easier to stamp because it is more ductile, thereby increasing the die life. On the other hand the regular 4 per cent silicon steels have always given a lower watt loss but the new process which has been developed can produce on a commercial basis coils not having a watt loss in excess of 0.45 watts per pound, the Epstein test strips all being in the direction of rolling. In this material the watt losses and permeability in the direction of roll-

ing are better than that at right angles to direction of rolling in the present state of the development.

In Fig. 6 the torsion magnetization curve of this new silicon steel development is shown. When compared with the torsion curve of the single crystal it will be seen that they nearly coincide, yet the grains in this particular material were smaller than those of the material whose magnetization curve is shown in Fig. 5.

It would be a long story to explain in detail how this new material was developed, and irrelevant to the subject of this paper,

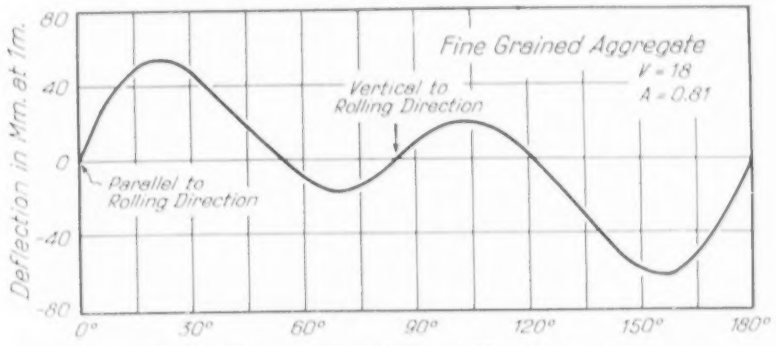


Fig. 5—Torsion Magnetization Curve of Fine-Grained Aggregate.

but the methods which must be followed in order to obtain these results will be discussed in some detail. Hundreds of specimens were prepared in various ways and the magnetic moments recorded and many of them X-rayed. The methods of cold-rolling and heat treating were carefully recorded for each specimen and the effect on the magnetic moment recorded. These experiments soon showed that only certain heat treatments and cold rolling would produce the desirable results, especially where one intermediate heat treatment was given; that is, in order to get magnetic moments approaching that of single crystals certain precautions are necessary both in the cold rolling practice and especially in the heat treatment of the hot and cold-rolled materials.

In the very beginning these experiments showed that the hot-rolled strip would play an important part in developing a better silicon steel in the form of heat treated cold-rolled strip. In some of the processed hot-rolled strips a high magnetic moment was realized. In other cases performing the experiments in exactly the same way a much lower magnetic moment was developed in the final product. Continued experiment showed that the hot-rolled strips could only be made in a limited number of ways so as to be suitable

for the new process. The type of hot mill on which the strip was rolled determined the heating and finishing temperature.

The hot-rolled strips which gave very good results and a group which gave inferior results were studied with the torsion magnet and it soon became evident that hot-rolled strip suitable for the process was characterized by a limited number of torsion curves both in regards to magnitude and to symmetry. The slabs are heated, rolled and finished at a temperature which, when tested in the torsion magnet, will fall within the limits allotted by previous tests on hot-rolled strip which gave good results.

The proper heat treatment of the hot-rolled strip in order to obtain the ultimate results in the final product had to be investigated. It was soon found that the heating temperatures best suited for a 3 per cent silicon steel were around 1600 degrees Fahr. In some instances it was possible to either box anneal the material by giving it a fairly rapid heating and holding it at the temperature and then giving it the regular slow cooling or it could be heat treated continuously in a continuous electric roller hearth furnace. In general it was found that the best magnetic properties could only be developed when the hot-rolled strip was heat treated in the continuous furnace in which the rate of heating was rapid and held only a few minutes at the temperature and then cooled rapidly. The rate of cooling is only a matter of convenience rather than an essential characteristic of the process. After the heat treating operation it was found that if the material was cold-rolled directly to gage and given the final heat treatment at a temperature in excess of that of annealing temperature of the hot-rolled strip only a fair magnetic moment could be developed in the final product. Again if the hot-rolled strip is cold-rolled directly to gage and given a final heat treatment at the same temperature a much inferior magnetic material was developed. The best results are found when the hot-rolled strip which has the proper characteristics suitable for this process, is heat treated and cold-rolled to an intermediate gage and then heat treated at a predetermined temperature and then cold-rolled to final gage and again heat treated at a temperature around 2000 degrees Fahr. The rate of heating and the time at temperature are of great importance in controlling the permeability and watt losses and the grain size. The essential features of the process are to start with the suitable hot-rolled strip and then by using the proper procedure it is possible to produce a fine or small grained material having the mag-

netic moments approaching that of a single crystal. This is accomplished by the proper distribution of cold rolling and heat treatment.

It has been found by thousands of experiments on the torsion magnet that the best intermediate gage varies depending upon thickness of the hot strip, its heat treating temperature, the chemical composition, the grain structure of the hot-rolled strip and temperature and the manner in which the hot-rolled strip was processed, etc. In general the best intermediate cold rolling gage lies somewhere in

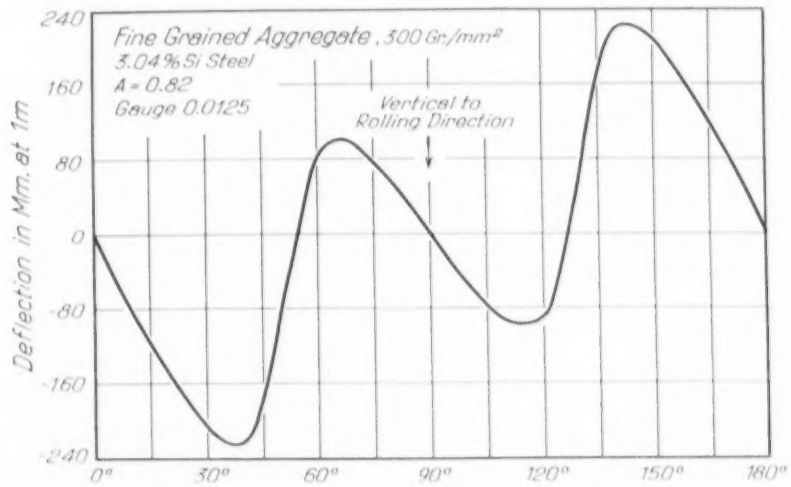


Fig. 6—Torsion Magnetization Curve of Fine Grained Aggregate. (300 Grains Per Square Millimeter.)

between 0.025 to 0.040 inch. One of the most important steps in the whole process, if an outstanding result is to be obtained, is in the intermediate heat treatment. Results have shown that the intermediate heat treatment must be rapid and that the strip must only be held at the temperature a limited number of seconds in order to get complete recrystallization and keep preferred orientation and grain distortion at a minimum. This is easily accomplished because the annealed hot strip is given a rather heavy reduction before this heat treatment is given in which the reduction varies between 50 and 80 per cent. For a 3 per cent silicon steel the intermediate heat treatment is usually from 1600 to 1800 degrees Fahr., after which the heat treated strip is given a further cold rolling to gage. Assuming that 0.032 inch was found to be best place to stop the initial cold rolling the best final gage in general will be approximately 0.0125 inch. Before the final heat treatment is given all of the samples of cold-rolled strips ready for the final heat treatment are run through the furnace at various speeds which are known to have given best

results in the past. Moment tests are made of these strips and the speeds which give the best results are then applied to the various coils.

When the hot-rolled strip is cold-rolled to final gage without any heat treatments and tested in the torsion magnet it is invariably characterized by a high magnetic moment. When it is given the final heat treatment the magnetic moment is always lowered a large amount. This is a very strange phenomenon. An X-ray investigation was made of these cold-rolled strips prior to the final anneal, one group of which was characterized by a low magnetic moment and the other by a high magnetic moment. The X-ray did not reveal any difference in the two structures.

After the final cold rolling the strip is heat treated. The time at temperature is of great importance. Recently it was found that if the strip was held too long in the continuous roller hearth furnace the magnetic moments were lower and the grains were relatively large and the watt loss was in excess of 0.55 watts per pound. However, when the same strip was run through the furnace at a much greater speed a very fine-grained silicon steel resulted, characterized by a very high magnetic moment and the watt losses were under 0.46 watts per pound. This immediately illustrates how important it is to determine the rate of speed the strip should be run through the furnace for the final heat treatment. Again some strips must be run at a slower speed in order to get good results. In general, we usually heat treat the final strip at about 2000 degrees Fahr. in an electric roller hearth furnace. When made in this way the strip is free from coil set, the surface is absolutely free of pits and is very smooth and due to the accuracy to which the 4-high single stand mill can cold roll, the stacking factor will be very high due to the flat surface. This new material is very ductile and will stand as high as 40 bends before failure which is far in excess of present day silicon steel made by the regular hot sheet method.

In Fig. 7 is shown the magnetization curve of this material in which the maximum permeability is 22,000 but permeabilities as high as 27,000 have already been reached. The permeability at 16,000-B is in excess of 2200 and we have already obtained in excess of 2500, which is about 6 to 8 times that of the present day silicon steel and exceeds that of Ruder's[6] single crystal silicon steel strips. The

[6]W. E. Ruder, "Magnetization and Crystal Orientation," TRANSACTIONS, American Society for Steel Treating, Vol. 8, July 1925, p. 25.

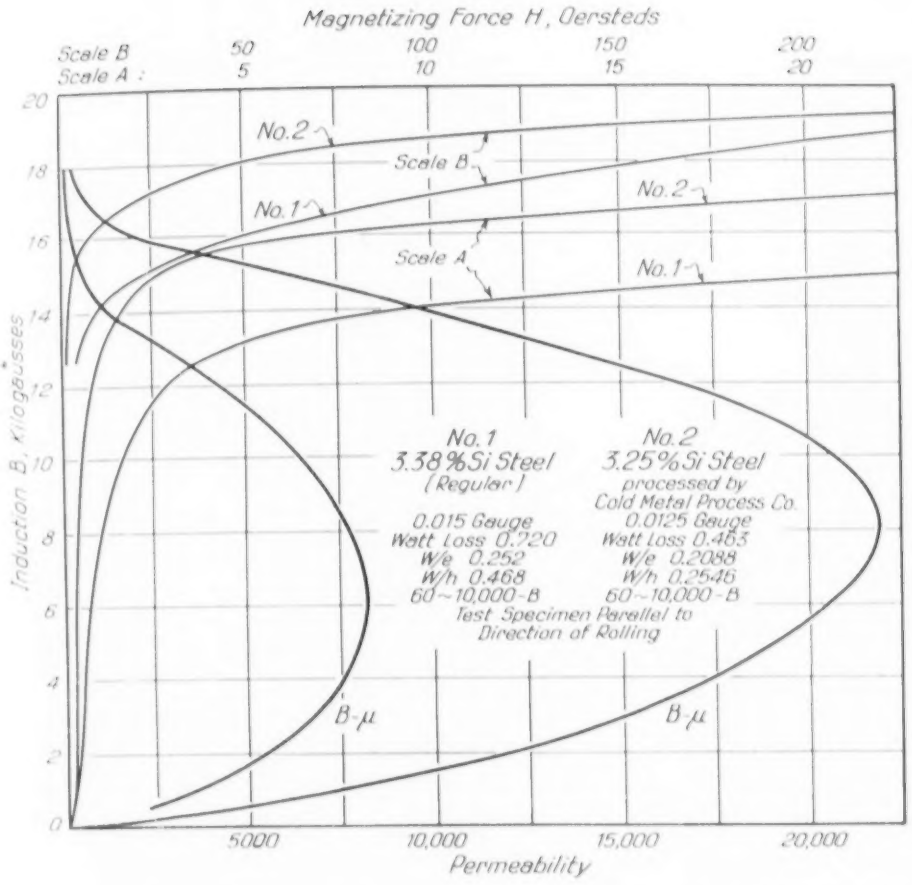


Fig. 7—Permeability and Magnetization Curves Comparing New Product with that of Present Day Hot Rolled Sheet.

watt loss of this particular strip was 0.46 watts per pound when measured at 60 cycles, 10,000-B, all the strips being cut in the direction of rolling.

After the final heat treatment in the continuous treating furnace the material is cut to the proper size for use in electrical machinery, after which it must be given a stabilizing heat treatment; this usually improves the electrical and magnetic characteristics. The watt losses are sometimes lowered or have remained the same but in general the permeability at the higher inductions has been markedly increased after the stabilizing heat treatment. If this treatment is not given to the material it will, as a rule, age; however, when the proper stabilizing treatment is given no aging occurs. In general, the best magnetic properties are usually found in the materials which are about 0.010 to 0.014 inch thick.

Ruder points out that material 0.010 to 0.12 inch in thickness can be handled economically in the building of electrical machinery without increasing the manufacturing cost. Likewise it is possible to cold roll silicon steel to almost any gage on the 4-high single stand cold rolling mills without increasing the cost of rolling. This type of mill also has the advantage in that the gage can be accurately maintained to within 0.00025 inch and the shape of the strip can be kept remarkably flat. This naturally affects the stacking factor and justifies a thinner gage provided the electrical and magnetic properties are superior in all respects to the thicker material.

This material can also be supplied in gages heavier than 0.010 to 0.015 inch but so far the results in general are not as good as that of the thinner material; however, heavier gages having the same properties as that of the thinner material may be possible in the future.

In 1914 Ball and Ruder[7] pointed out that the permeability should be better in the direction of rolling because the heat treatment never entirely wipes out the elongation of the grains. They did not realize that magnetic moments actually exist in the sheet in which the component parallel to the rolling direction is always much greater than at right angles. This and not the elongation of the grains explains the difference in permeability with and cross grain.

Spooner[8] points out that whenever it is possible silicon steel should not be used with a magnetic induction perpendicular to the direction of rolling since the properties are not so good. In the case of this new material the results are so much better in the direction of rolling that wherever it is possible the suggestions of Mr. Spooner should be carried out.

In the past it has been believed that as the grain size of silicon steel increased, the watt losses became less and the permeability higher but in this new material the reverse is true; as the grain size becomes smaller the watt losses decrease and the permeability increases. Messkin and Kussman in their book on "Ferro-Magnetic Alloys and Their Commercial Application," Julius Springer, Berlin, 1932, after having made a complete survey found that consensus of opinion was that the coercive force and the hysteresis loss are definitely diminished by increased grain size.

It has already been pointed out above that the X-ray method of

[7] J. D. Ball and W. E. Ruder, "Eddy Current Losses in Sheet Steel as a Function of Resistivity," *General Electric Review*, Vol. 17, 1914, p. 487-81.

[8] Thomas Spooner, "Properties and Testing of Magnetic Materials," p. 116, McGraw-Hill Book Co., Inc., New York City.

crystal analysis as applied to metallurgical problems was almost indispensable in this investigation. However, at this time it would hardly be possible to present the entire X-ray data collected during the past three years on the development of this new magnetic material. In the near future a separate paper dealing with the X-ray investigation will be presented. At this time, however, a few of the more interesting X-ray results will be presented.

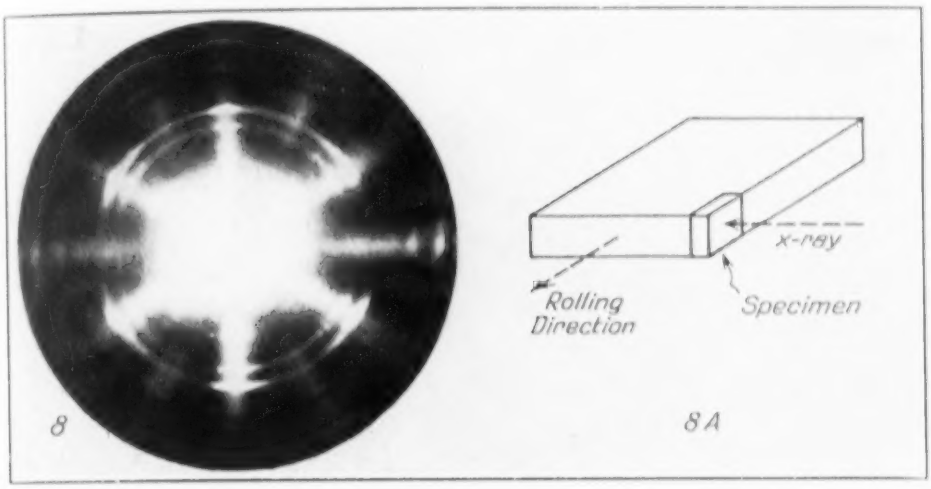


Fig. 8—X-Ray Crystal Structure of Hot-Rolled Strip. Direction of Rolling Corresponds to the Lengthwise Direction of this Page.

Fig. 8a—Sketch Showing Manner of Selecting Specimen and Direction of Rolling and X-Rays.

In Fig. 8 the crystal structure of the hot-rolled strip is shown. The X-ray beam was transmitted through a specimen which was cut parallel to the direction of rolling and at right angles to the surface. In other words, the X-ray beam was transmitted in a transverse direction with respect to the direction of rolling and parallel to the surface. It is very interesting to see the intense fiber structure which is developed even in hot rolling. This is not what might be called the most typical of the hot-rolled structures but was presented because of the very sharply defined fiber structure. This material in the final stage usually gives a very excellent result from magnetic point of view after the proper processing has been applied to it. Other specimens of the hot-rolled strip have been prepared in which the X-ray beam was transmitted in a direction at right angles to the surface and in which case a nearly random grain structure was typical. If the X-ray beam is transmitted in a direction parallel to the direc-

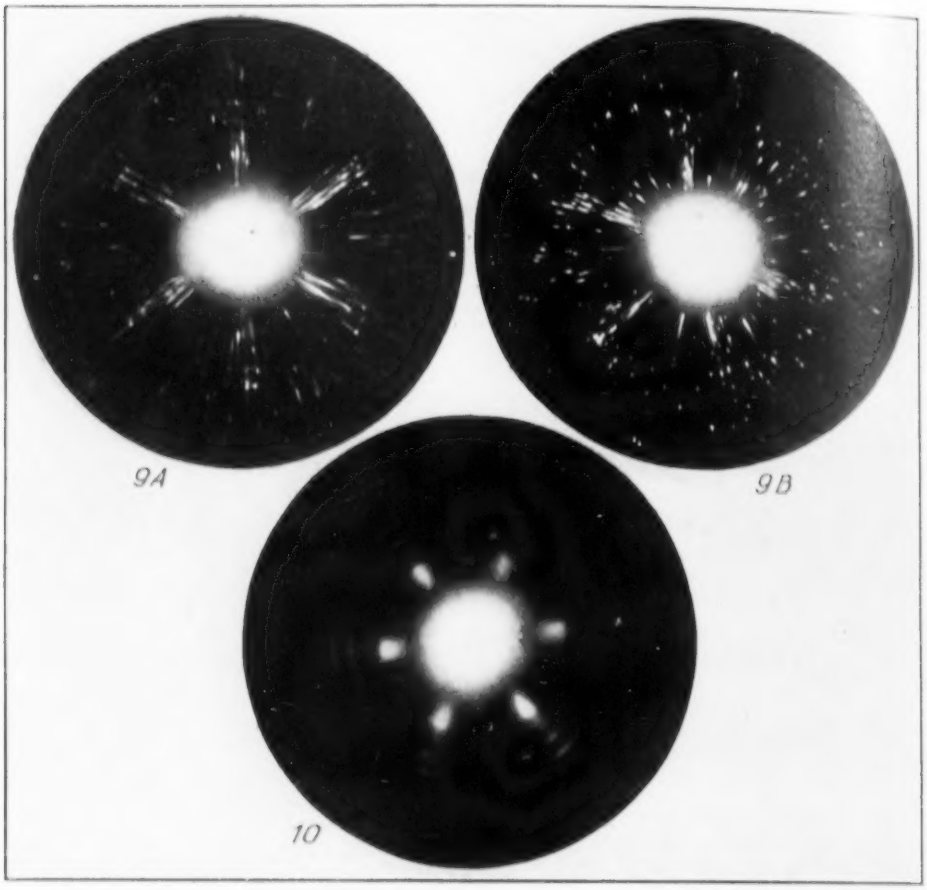
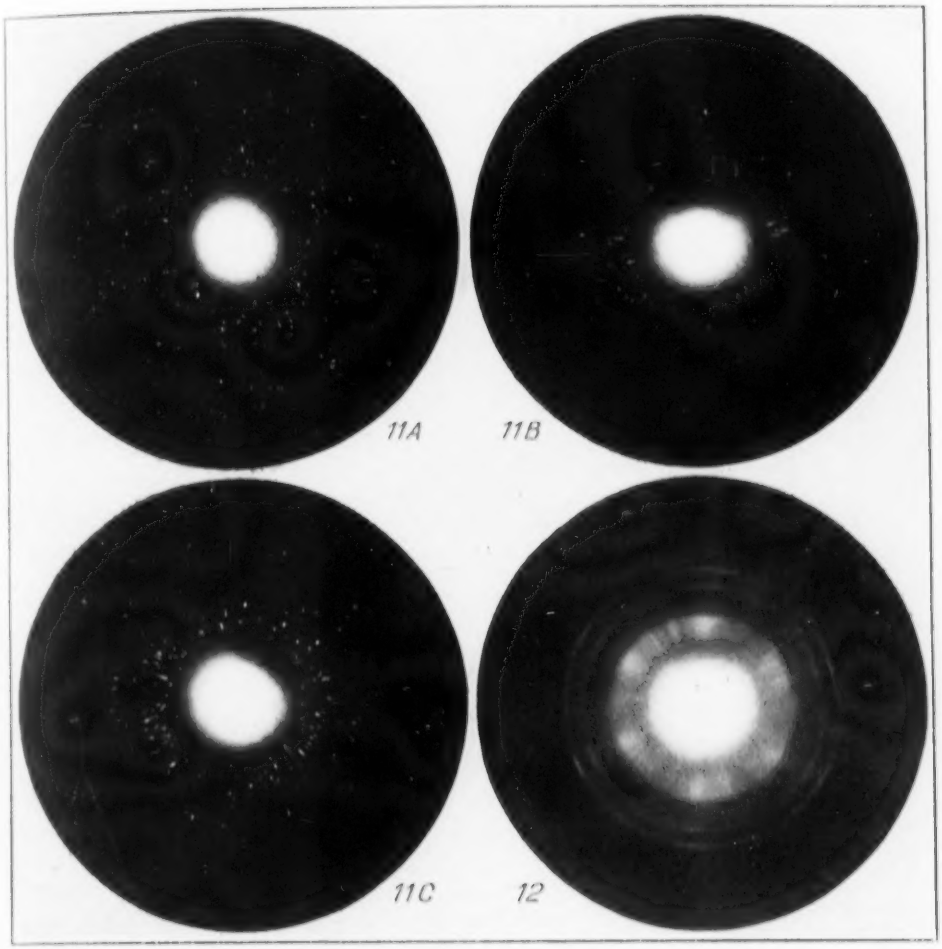


Fig. 9A and 9B—Typical Heat Treated Structures of Hot-Rolled Strip X-Rayed as Shown in Fig. 8a.

Fig. 10—X-Ray Laue Diagram of the Hot-Rolled Strip Cold-Rolled to the Intermediate Gage 0.100-0.030 Inch. The X-Ray Beam was Transmitted as Shown in Fig. 8a.

tion of rolling a very weak fiber structure appears. In general, however, no marked orientation is found.

In Fig. 9 we show the structure of the heat treated hot-rolled strip. In this particular instance the grains have been nearly completely recrystallized and the directional properties are almost entirely removed. However, after a complete survey on about 200 hot-rolled strips made in numerous ways and heat treated in various ways it was found that in order to get good results in the final product complete recrystallization is not absolutely necessary. The preferred orientation need not be entirely eliminated but the recrystallization must be as rapid as possible. After the hot-rolled strip has been heat treated it is subjected to cold rolling which is stopped at an intermediate gage.



Figs. 11A, 11B, 11C—X-Ray Laue Diagrams of the Heat Treated Cold-Rolled Strip of Fig. 10. These Structures all gave Good Results in the Final Strip.

Fig. 12—X-Ray Diagram of the Heat Treated Cold-Rolled Strip, Cold-Rolled to the Final Gage Prior to the Final Heat Treatment. X-Ray Beam was Transmitted Perpendicular to the Surface.

In Fig. 10 the transverse structure of this cold-rolled strip is shown. It is surprising to see how intense the fiber structure is, and in every specimen examined in this way a very intense fiber structure has always been found. However, when the X-ray beam was transmitted at right angles to the surface a very weak fiber structure is usually found, but there are exceptions. When the cold-rolled strip is now heat treated this transverse fiber structure is removed and the grains are completely recrystallized, although in general they lack perfection of crystallinity. Under ideal conditions such strips in the heat treated state should be characterized by a complete recrystallization free of grain distortion and uniform in size and

the recrystallization must be rapid. Such structures have actually been obtained in practice, although the recrystallization may be complete some grain distortion is found after heat treatment, see Fig. 11. The material after intermediate heat treatment is then further cold-rolled and the X-ray pattern of the material is shown in Fig. 12 and its heat treated structure is shown in Fig. 13. It will

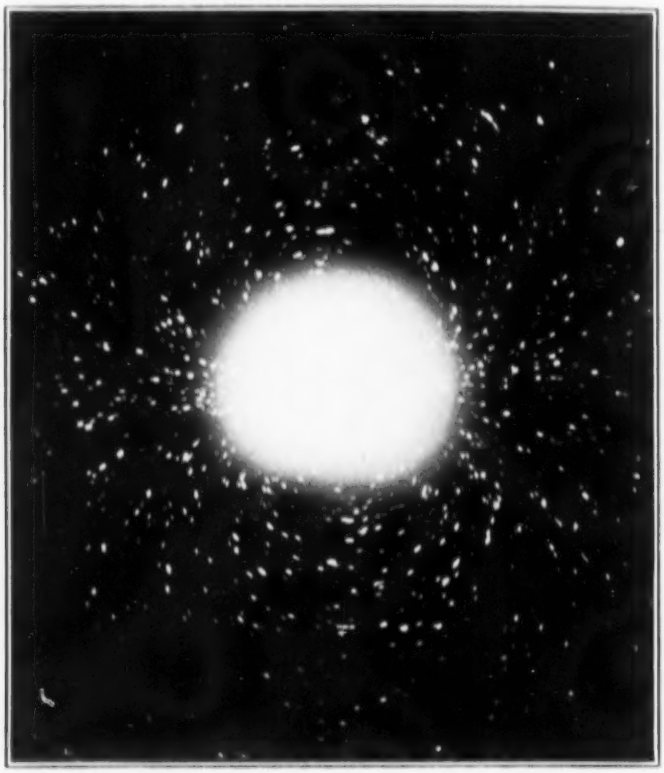


Fig. 13—X-Ray Laue Diagram of the Finished Strip, 0.0125 Inch Thick. The Specimen was Moved at Regular Intervals During Exposure.

be seen that in the final heat treated condition the grains are distributed at random and are nearly perfect in structure. In this particular X-ray diagram of the heat treated structure the specimen was moved parallel to itself during exposure to the X-ray in order to bring as many grains into the proper reflection position as possible, yet no trace of preferred orientation is in evidence. We have made over one hundred X-ray diagrams of the various silicon steel strips in this way and in not a single instance did we find any preferred orientation. It might be well worth noting that although the crystal structure is most important, structures have been found in

material which had a relatively small moment and naturally the magnetic properties were very poor but in combination; that is, in material characterized by a high magnetic moment in which the grain structure is characterized by a random grain distribution of uniform size and the grain size being small and as nearly perfect as possible the best magnetic properties are always found.

The X-ray has been of great value in following the effect of time of heat treatment on the strips in the various stages of processing. In general, we have found that in every case where the final materials exhibited a high magnetic moment and those in which the crystal structure was well developed gave the best results. With these facts in mind we will continue to improve the product by controlling the process from the open-hearth to the finished product in order to obtain as perfect a crystal structure as possible to produce a fine-grained aggregate in which the magnetic moment approaches that of a single crystal.

In the present paper only one preferred practice has been described in some detail. For those interested in the variations in the procedure complete details will be found in U. S. Patent 1,965,559.

The author wishes to express his appreciation to H. S. Lamb for his assistance in developing the magnetometer used in these tests.

DISCUSSION

Written Discussion: By Jacob Kunz, professor of physics, University of Illinois, Urbana, Illinois.

The results given in this paper are as important as they are surprising. It has so far been known that high permeability is connected with single crystals of very pure metals. Here on the contrary we have high permeability of a steel containing over 3 per cent silicon in a fine-grained crystalline aggregate. The conditions under which this material has been obtained are indeed very exacting, though commercially possible. The temperatures have to be chosen in well defined intervals, the time of exposure to high temperature, and the time of cooling require only a few minutes. The hot-rolled strip of proper characteristics is heat treated, then cold-rolled to an intermediate gage, then heat treated, then cold-rolled to the final gage, and again heat treated at a temperature around 2000 degrees Fahr. (1150 degrees Cent.). Much depends on the intermediate heat treatment, which must be rapid, in order to get complete recrystallization and reduce preferred orientation and grain distortion to a minimum. The best magnetic properties, i.e., high permeability and low hysteresis loss, are obtained with fine grains of random distribution free of distortion and of uniform size. Hot-rolled strips show intense fiber structure in the X-ray photographs. Cold rolling apparently destroys this fiber structure, and recrystallization takes place in the intermediate and final heat treatment.

With respect to the crystalline aggregate there arise important questions which have not yet been answered. For instance, if two perfect single crystals were joined along a certain surface, how would this intermediate surface affect the magnetic properties? And if the two crystal surfaces meet at an angle, what is the nature of cementing material between the two surfaces? This material might be amorphous or crystalline. In the case of the iron-silicon steel of Goss, the magnetic properties of the elementary crystals do not differ much in different directions, so that if the elementary crystals form an aggregate, the magnetization in a given direction may be about the same as for a single crystal, especially if the material between the crystals is crystallized also. Moreover, the question arises as to the role of silicon. It may serve only to purify the iron from oxygen or it may enter into the gratings of the crystals.

Mr. Goss has given only the measurements of the perpendicular component of magnetization. He has undoubtedly measured the parallel component also. It is fortunate that the perpendicular component is so characteristic for the resultant properties of the material. Figs. 5 and 6 of Goss's paper show that the perpendicular component of magnetization in the direction of rolling is zero. It will be interesting to know whether the parallel component in that direction is a maximum or whether it is approximately constant for every direction in the plane of the sheet.

Roentgen rays tell us very little about magnetic properties. But recently Bradley and Rodgers[1] have shown that an annealed and slowly cooled Heusler alloy is nonmagnetic, but on quenching from 800 degrees Cent. (1470 degrees Fahr.) it becomes strongly ferromagnetic. At the same time the structure changes into an entirely body-centered cubic, with a face-centered superlattice.

I wish to congratulate Mr. Goss for the important contribution made with the simple methods described in his paper.

Written Discussion: By S. R. Williams, professor of physics, Amherst College, Amherst, Mass.

The interesting work of Mr. Goss has opened a new field of investigation. It has completed the overthrow of our ideas about the orientation of crystal structure as an all important factor in high permeabilities.

The work of the two Comptons[2] and of Barnett[3] has definitely shown that in the process of magnetization it is not the crystal which is rotated, nor even the atoms themselves. If the idea of something being rotated has any validity in the picture of what goes on when a body is magnetized, then it must be referred to the orientation of a spinning electron as the elementary magnet.

Mr. Goss has gone a step farther, he has shown that the exceptional magnetic properties of the single crystal is not necessarily due to its configuration, but rather due to the arrangement, location and orientation of the spinning electrons themselves. If the splendid magnetic properties found by Mr. Goss

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[1]A. J. Bradley and J. W. Rodgers, *Proceedings*, Royal Society of London, Vol. 144, 1934, p. 340.

[2]Compton and Trousdale, *Physical Review*, Vol. 5, 1915, p. 315.
Compton and Rognley, *Physical Review*, Vol. 16, 1920, p. 464.
Compton and Rognley, *Science*, Vol. 46, 1917, p. 415.
Compton and Rognley, *Philosophical Magazine*, Vol. 41, 1921, p. 279.
Compton and Rognley, *Journal*, Franklin Institute, Vol. 192, 1921, p. 145.

[3]Barnett, *Proceedings*, American Academy of Arts and Sciences, Vol. 60, 1925, p. 127.

are due to the orientation of the spinning electrons within the crystals, then the crystal structure does not control this orientation.

One cannot easily picture the heat treatment as the cause of the preferred orientation of the spinning electrons. The conclusion seems to be that the rolling process is the chief factor which determines the magnetic properties found by Mr. Goss.

There are several lines of investigation which should be followed. The rolling and heating processes may give random distribution to the crystals, but it does not preclude a preferred orientation of a series of planes of cleavage (joints) between the crystals. The rolling process accentuates these slip planes, which in turn affects the hardness of the material. Mechanical hardness is most intimately associated with the magnetic properties of materials.[4] These relations call for a comprehensive study.

Also the effects of longitudinal and transverse joints in ferromagnetic substances should be investigated more thoroughly. They play a most important part in the behavior of metals.[5]

Furthermore, in the theories of hardness much is said about the material between the slip planes.[6] What are its magnetic properties?

Mr. Goss is to be congratulated on this important contribution which he has made to the subject of magnetic theories.

Written Discussion: By W. E. Ruder, General Electric Company, Schenectady, N. Y.

As the author has stated, the idea of cold rolling silicon steel for magnetic applications is not new. There has long been a desire on the part of users to obtain this product in coils for particular applications where the stamping operation might be speeded up, but up until recently no extended use of strip has been made, not because of inferior quality as stated, but rather because of the higher cost of cold-rolled strip as compared to hot-rolled sheets of equal magnetic quality. The author's statement regarding the production of a "greatly improved silicon steel" by means of the cold rolling method requires some modification if we are to use his values of 0.45 watts per pound Epstein test on all "with the grain" samples, as a standard of comparison. Hot-rolled sheets of equal and better quality than this, when compared on the same basis, have been in use for some years now. Since total loss values are also dependent upon thickness as well as rolling direction, direct comparison can only be made with material of the same thickness. For example, the value 0.54 watts per pound quoted from Pomp (S. & E. Vol. 44, 1924) was obtained on 0.0078-inch and cannot, therefore, be compared directly with the watt loss figures given in the paper.

In Fig. 7, comparing a hot-rolled with a cold-rolled material of nearly the same silicon content, the eddy current loss in the cold-rolled sample, because

[4]Williams, *Science*, Vol. 65, 1927, p. 306.
[4]Williams, *Science*, Vol. 66, 1927, p. 358.
[4]Williams, TRANSACTIONS, American Society for Steel Treating, Vol. 14, 1933, p. 741.

[5]Ewing and Low, *Philosophical Magazine*, Vol. 26, 1888, p. 274.
Maurain, *Journal de Physics*, Vol. 1, 1902, p. 151.
Williams, *Physical Review*, Vol. 33, 1911, p. 60.
Williams, J. O. S. A., Vol. 10, 1925, p. 109.
McKeehan, J. O. S. A., Vol. 11, 1925, p. 169.

[6]Jeffries and Archer, "The Science of Metals," Chapter 4, 1924, p. 63.

it is thinner, is 0.043 watts per pound less than that of the hot-rolled sample, which amount, if added to the total loss of the cold-rolled sample, would make it 0.506 instead of 0.463.

The unusual shape of the magnetization curve produced by cold rolling (Fig. 7), approaching that of a single crystal, appears to be a distinct characteristic of cold-rolled polycrystalline magnetic material. Smith, Garnett, and Randall (British Patent '1930') (U. S. Patent 1,915,766 '1933') showed that by selecting the proper degree of cold reduction and combining it with a suitable heat treatment, a very much improved permeability could be obtained in magnetic materials of various kinds and explained their results by pointing out that the high degree of preferred orientation so developed is one most favorable to magnetization in the direction of rolling. Our own experiments on cold rolling silicon steels containing from 3 to 5 per cent silicon confirmed this, although at that time we had not exceeded cold reductions of 55 per cent. In this paper, however, the author speaks of avoiding preferred orientation for best results and deduces from his X-ray studies that he has no trace of preferred orientation. He explains his results by a "magnetic moment" developed by cold rolling. This "magnetic moment" test is in reality nothing more than a simplified permeability test used by many investigators to determine directional properties, which in turn must be caused by some change in the metal produced by the cold rolling. Our own experiments show that above a certain degree of cold reduction the alloy tends to recrystallize, after annealing, into grains having decidedly similar orientations. Sixty to seventy per cent of the grains will have 1 1 0 planes parallel with the surface and 1 0 0 axes varying only a few degrees from parallel with the rolling direction. The remaining 30 to 40 per cent have 1 0 0 planes parallel to the surface and 1 0 0 axes parallel to the rolling direction. From our previous work on the effect of orientation on magnetization, referred to by the author, we find that the directional properties of cold-rolled strip check very closely with what we would expect from these orientations.

If such a preferred orientation exists, why does not the X-ray show it up? We have made many X-ray examinations with results similar to those given in this paper. Cold-rolled specimens show decided orientation which subsequent heat treatment apparently obliterates. Optical methods show that the preferred orientation is still there. The only explanation we have to offer at this time is that because of the large grain-size characteristic of annealed silicon alloys, the X-ray method used is not reliable. The author refers in several places to the very fine grain structure obtained. Examination of the structure of several strips which he has sent out as samples showed them to be of irregular but fairly large grain-size—in most cases of the same order as that of good hot-rolled sheets.

In view of this close agreement we have found between the preferred orientation resulting from proper cold rolling and heat treatment and the properties of similarly oriented single crystals, we see no "radical departure from present day theories of ferromagnetic properties."

While the author does make some reference to directional properties, particularly about cross and with "grain," we feel that this needs a little more emphasis, particularly from a practical view point. A standard Epstein sample

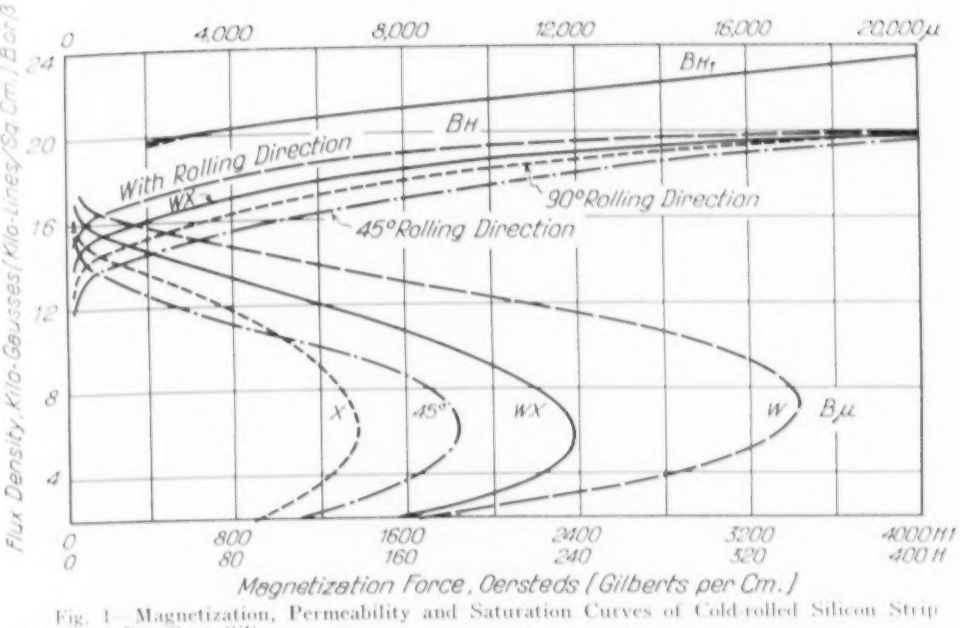


Fig. 1—Magnetization, Permeability and Saturation Curves of Cold-rolled Silicon Strip Steel 2.96 Per Cent Silicon.

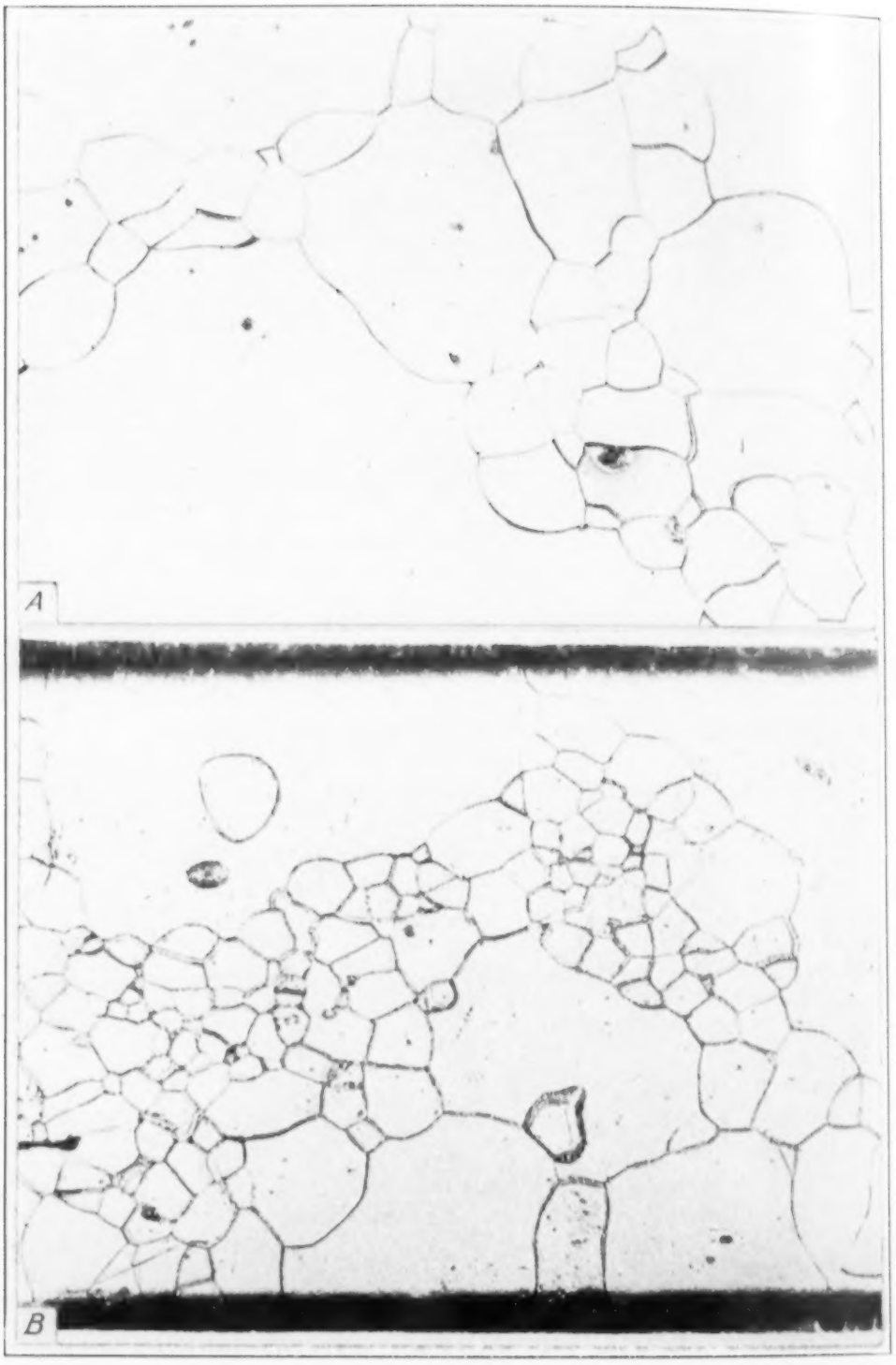
calls for half the punchings cut with and half across the rolling direction, and experience has shown that this is a representative test for hot-rolled sheets and checks fairly well with a ring test. With cold-rolled strip, however, this is not so, since, because of its peculiar preferred orientation, the 45-degree direction is poorest, i.e., has lowest magnetization and highest loss. (See Fig. 1 of this discussion).

In hot-rolled sheets directional properties also exist, and in individual samples these are quite marked, but because of the variations existing in pack rolling, these vary considerably from specimen to specimen with the result that in fairly large lots, such as may be represented by the Epstein sample as compared with a single disk, and using half with and half cross rolling direction strips, these are fairly well averaged out. In spite of these variations, two types of preferred orientation do prevail in about equal percentages; namely, a 1 0 0 plane parallel to the surface with a 1 1 0 axis parallel to the rolling direction, and a 1 1 0 plane parallel to the surface with a 1 0 0 plane parallel to the rolling direction.

As to the references made to the effect of grain-size, we have gone into this in some detail in a paper to be presented later at this meeting. Reference is made to our 1914 paper where we gave the elongation of grains as a possible explanation for directional properties. It was not until about three years later that we discovered the important effect of orientation. The best permeability values we have obtained for single crystals was a maximum value of 24000 and 3000 at B = 16,000. This is for a crystal having an orientation still considerably off a 1 0 0 face perpendicular to the direction of the flux. (See Table XI by Ruder).*

Maximum permeability is not, as the author states, markedly reduced by

*W. E. Ruder, "Influence of Grain-Size on Magnetic Properties," TRANSACTIONS, American Society for Metals, Vol. 22, No. 10, 1934, p. 1120.



A—Photomicrograph of Face of Goss Cold-rolled Strip. × 250.
B—Photomicrograph of Cross Section of Goss Cold-rolled Strip. × 250.

silicon. On the contrary, increased silicon increases the maximum permeability of commercial silicon steels.

Written Discussion: By N. A. Ziegler, Power Use Research Department, The West Penn Electric Co., Pittsburgh.

Mr. Goss should be congratulated on his very interesting paper. His description of the manufacturing procedure and properties of the new magnetic material is very clear, and it only remains to be wished that he would give an explanation as to what happens inside of his material, which is responsible for these good properties.

Our belief so far was that the regularity of the space lattice is the principal factor which determines the maxium permeability and hysteresis loss of magnetic materials. This regularity of the space lattice could be destroyed, with a consequent effect on the magnetic properties, by one of the following three factors: (1) Mechanical distortion; (2) Distortion at the grain boundaries; (3) Distortion produced by the foreign elements, "impurities." So far the experimental evidence has been in favor of this opinion, and for this reason our efforts have been concentrated on eliminating these three factors, in order to produce improved magnetic materials.

It would be interesting to know what methods and means Mr. Goss used in controlling the chemical composition of his samples. It is known that variation of carbon and sulphur content in silicon iron, even within thousandths of one per cent, is liable to produce variations in magnetic properties of the same order of magnitude as he describes.

Oral Discussion

Dr. Kent R. Van Horn:[7] Before condemning the X-ray in general, it might be advisable to determine if the various methods of analysis have been exhaustively applied, and also if the most has been obtained from the diffraction patterns. I have reference to page 518 in which the author describes a number of photograms of strip steel .005 inch in thickness in which the X-ray beam had been directed perpendicular to the rolling plane of a stationary specimen. There were no patterns prepared with the X-ray beam parallel to the rolling plane and either perpendicular or parallel to the rolling direction. In addition there has been no attempt to rotate the specimens during exposure.

This criticism is not intended specially for the author, as nearly all the preferred orientation investigations in Germany and in this country, until a few years ago, procured results by examining thin sheets perpendicular to the surface or rolling plane. Later, the work progressed so that three directions, at right angles to each other (one perpendicular and two parallel to the sheet surface) were exposed, thereby making it possible to more easily reveal preferred orientation. It is possible when a fiber structure is present to direct the beam parallel to the fiber axis thereby procuring a random film, but an examination in the other directions would detect the preferred grain arrangement. Consequently, I think that Mr. Goss's remarks on page 519, where it is stated that no preferred orientation exists, should not be as emphatic unless these other examinations had been made. The possibilities of the X-ray method

[7]Aluminum Co., of America, Cleveland.

certainly have not been exhausted, and it should be necessary to make these exposures and, more important, to employ the new pole figure method where the sheet is examined at different angles, perhaps every 5 degrees, thereby giving a more complete story.

Thusfar my remarks have been confined to manipulation and technique, but this problem has had some serious practical significance. A few years ago the industrial field was startled by the introduction of a special rolling mill which was claimed to be so unique that no preferred orientation was produced by large reductions. This discovery caused all the large fabricating industries to initiate X-ray investigations on this new rolled metal with the universal result that the sheet rolled by this mill had as much preferred grain arrangement as any of the common two-high mills. The error was found to be attributable to the inadequate technique mentioned in the preceding paragraph. The sheet produced from the mill was not examined in any direction except perpendicular to the surface and the rolling direction, and no attempt was made to make a pole figure or some more exhaustive study.

In this connection, it is interesting to note that Mr. Goss obtained a similar conclusion which he describes in the present paper. However, I might add that this mill has a number of other advantages such as the ability to procure high reductions between anneals, accurate dimensions, etc., but it was a poor introduction for the mill, and also caused the X-ray method to be regarded with skepticism by those concerned. Consequently, when the X-ray investigations are employed, it is necessary to be certain that the possibilities of the method and the interpretations are exhausted. Then also, which is probably the case in this new transformer sheet steel, metallurgical transformations and phenomena are not always accompanied by changes detectable on X-ray films. We cannot expect too much of this new experimental and commercial tool, and likewise any other tool or method, but the valuable information thusfar procured by the X-ray justifies a fair trial.

W. E. RUDER: I just want to point out that my apparent criticism of the X-ray method is not a criticism but really a sincere question. The chairman's remarks are well taken. Here we have a case where everything points (that is the optical method and the magnetic properties) to a distinct preferred orientation which we feel that the X-ray ought to show up and it does not. That is why I bring it up as a question before this meeting.

The question of taking these pictures at different angles of rotation—we have tried that too. We have taken it at about four different angles, feeling that we must see preferred orientation, but these were only moderately successful; that is there is some indication of preferred orientation but nothing like the optical method shows actually exists.

R. H. SCHAEFER:[5] My first question concerns Fig. 4, page 518. This X-ray diagram supposedly shows that the crystals are oriented at random. However, closer examination of the photogram seems to indicate that the orientation may be preferred. This may be questionable, but the point is that other photograms should be taken; for example, 90 degrees rotation around the rolling direction.

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[5]Metals Research Laboratory, Pittsburgh.

On page 527 a statement is found (it refers to Fig. 8): "this is not what might be called the most typical of the hot-rolled structures but was presented because of the very sharply defined fiber structure, whereas, other specimens of the hot-rolled strip have been prepared in which the X-ray beam was transmitted in a direction at right angles to the surface and in which case a nearly random grain structure was typical." In other words, preferred orientation may not be pronounced in the particular plane in which the photogram was taken.

I would also like to point out, as the previous speaker pointed out, that the specimens, although labeled small grain, seem to be otherwise, not only from optical methods but from the very size of the spots in the photogram. While the specimen has been moved to remedy this condition, the fact remains that the photogram still does not have enough spots to tell very much about the presence of a small degree of preferred orientation.

R. S. DEAN:[9] This paper brings out very many interesting points bearing on the relation between the magnetic properties of single crystal and multicrystalline substances. In our work on magnetic separation and magnetic properties of minerals in the Bureau of Mines, we have had an opportunity to study this relationship under conditions where we can perhaps get at it more easily than in metals. We can take for example a single crystal of magnetite, and there is very little question that the crystal of magnetite is essentially a single crystal and we can literally philosophize with the hammer; that is we can hit it with the hammer and crush it down to any desired size and we have exactly the same material differing from the original crystal only in that it is now made up of many very small crystals. We think that those crystals are not appreciably distorted because of the very considerable brittleness of magnetite, but I do not want to get into a discussion with the X-ray people as to whether our X-ray evidence for that is worth anything or not. In any case, however, we have discovered certain very clear cut relations between the recompacted magnetite crystalline solid and the original one. The relation between what we might call internal surface or specific surface and coercive force is exactly linear. The coercive force of magnetite in the ordinary crystalline state or even ground down to 20-mesh and brought to a constant packing density may be two or three, something of that order; while if you grind that material and get it down to a few microns in size and recompact it you will find the coercive force is about 130 and the relationship throughout a series of crystal sizes between those two is such as to show very definitely that the coercive force is a definite function of the internal surface.

I have hesitated to use that word "Internal surface" because it sounds a little like white lamp black, but I see that in one of the May lectures of the Institute of Metals that term is used and defined in exactly the same way that I have used it. Now to apply that to steel may be a little questionable, but we are inclined to think that the coercive force of a metal measures definitely and distinctly its internal surface, not its grain size. There may be many discontinuities in the lattice which would constitute internal surface other than the grain size as measured either with the microscope or the X-ray. Those discontinuities might arise from glide planes or precipitation of impurities. I am not sure that precipitation of nonmetallic impurities would bring about that

[9] U. S. Bureau of Mines, Washington, D. C.

change in magnetic properties. I think it would not. Our results so far indicate that if you take this finely grained magnetite, say with a coercive force of 130 and put in an inert material such as finely ground silica, you get a gradual indication that as the electrical contact between the grains is broken the coercive force drops to zero. That point has not been fully established, but the indications are that a nonmetallic inclusion would not increase the coercive force of steel, whereas a metallic inclusion by increasing the internal surface would increase the coercive force. Other magnetic properties follow coercive force, but I would prefer to restrict the deduction to that because it seems to bear a very definite relation to surface.

Author's Closure

In reference to Dr. Van Horn's pertinent remarks concerning the procedure employed in the analysis of this material may I offer the following.

All X-ray specimens both in the heat treated and cold-rolled state were always examined in three directions, namely:

(1) Perpendicular to the surface.

(2) Parallel to the direction of rolling.

(3) In a transverse direction.

In some cases various samples were studied at various predetermined angles. This study was made in order to find out in which direction preferred orientation would be expected. After an extensive survey on many specimens I found that the strongest fiber structure occurred in specimens which were X-rayed in the transverse direction; whereas, in specimens X-rayed in the other directions just mentioned a weak fiber structure or completely random grain distribution resulted. For convenience in this paper I have proposed to call the transverse direction the principle direction and it is this direction with which I am mainly concerned in this investigation; that is, what effect various heat treatments have on the fiber structure developed during the course of cold rolling or hot rolling as the case might be. In the past it has been considered necessary to give any material a very large area reduction before the fiber structure is developed. However, in the case of silicon steels at least when X-rayed in the principle direction I always found that it only required an area reduction of 10 per cent to develop a very well defined fiber structure; whereas, in the other two directions not a single trace of fiber structure could be detected. Since the transverse direction was found to be the most important insofar as fiber structure in concerned then nearly all X-ray specimens were prepared so that the X-ray beam would pass through the specimen in what I call the principal direction shown in Fig. 8-a of the text.

Fig. 14 will illustrate this point; namely: that it requires very little cold work to develop a marked fiber structure in this material.

In Fig. 14-a is the X-ray fiber structure of the hot-rolled strip as received. This specimen was studied in the principle direction. The grains are oriented sharply in a very definite manner during hot rolling.

In Fig. 14-b the heat treated hot-rolled structure is shown. In this particular instance the hot rolling fiber structure was removed; however, this is not the general case as seen in Fig. 9-a of the text. The reason this particular material was chosen was because heat treatment removed the fiber structure

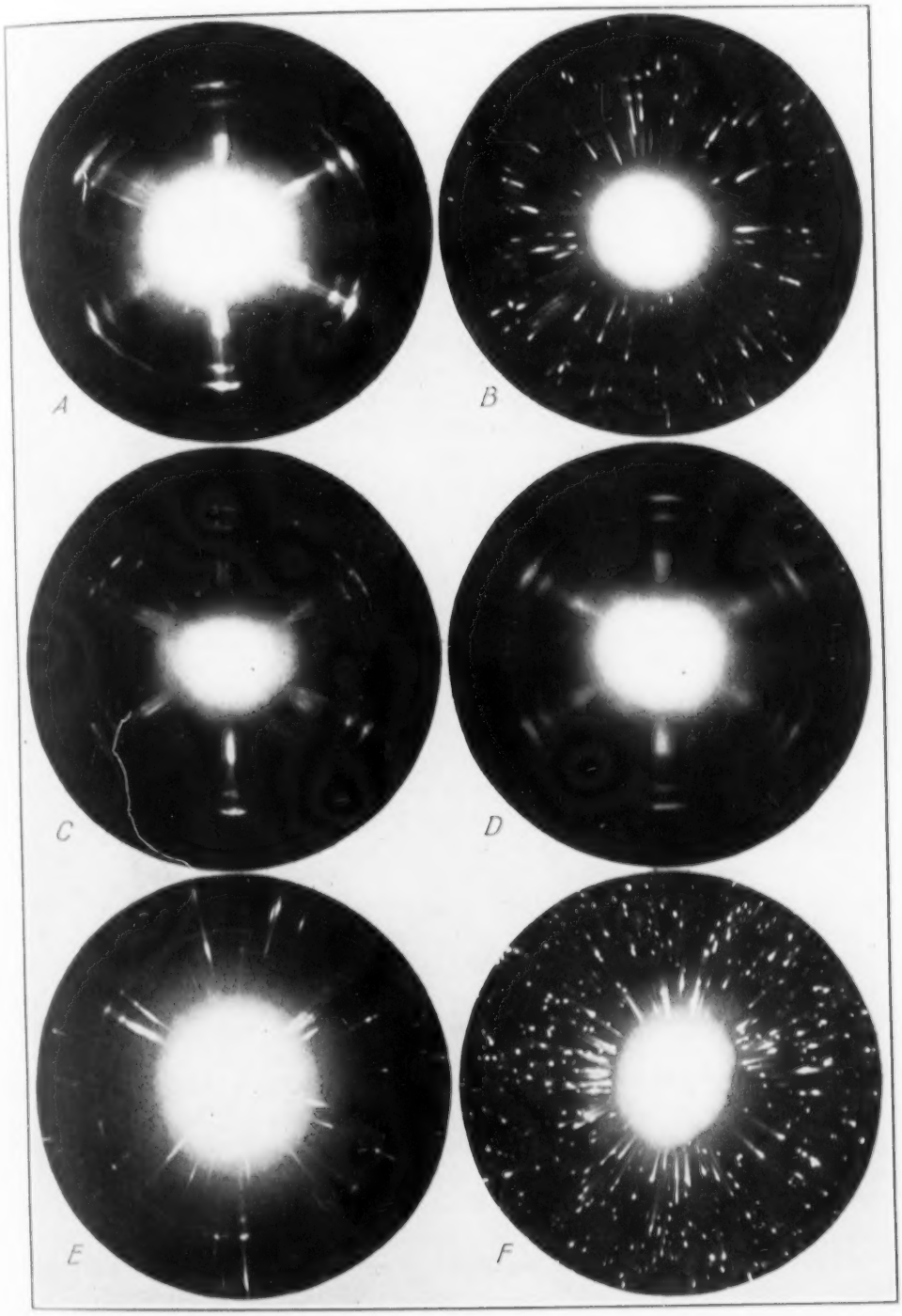

Fig. 14-a—X-ray Diffraction Pattern of a Hot-rolled Strip X-rayed in the Principal Direction.

Fig. 14-b—The Hot-rolled Structure after Heat Treatment, Again X-rayed in the Principal Direction and Showing Nearly Random Grain Distribution and Nearly Complete Recrystallization.

Fig. 14-c—The Heat Treated Hot Strip Given an Area Reduction of 10 Per Cent. X-rayed in the Principal Direction.

Fig. 14-d—The Hot-rolled Strip after an Area Reduction of 66 Per Cent.

Fig. 14-e—X-ray Diffraction Pattern of a Hot-rolled Strip X-rayed in the Principal Direction. Note the Strong Orientation of the Grains yet this Material had a Very Low Magnetic Moment.

Fig. 14-f—The Same Specimen X-rayed in the Direction Normal to the Sheet.

developed during hot rolling and at the same time completely recrystallized the hot-rolled structure. This material was then given an area reduction of 10 per cent and the specimen again X-rayed in the principal direction.

Fig. 14-c shows distinctly that an area reduction of only 10 per cent definitely orients the cold-worked fragments. This is certainly contrary to the opinion of many investigators who believe that a considerable amount of cold work is required in order to orientate the grains in some definite preferred direction. I found that further cold working only results in the development of a fiber structure as shown in Fig. 14-d in which the specimen had received an area reduction of 66 per cent. It is evident from this diffraction pattern that the crystals have been refined a great deal more than those in Fig. 14-c but the orientation in either case is the same.

I have made a complete study of various heat treated hot-rolled strips given area reductions ranging from 10 to 80 per cent in intervals of 10 per cent and this type of fiber structure is always developed without exception varying only in degree. When the fiber structure shown in Fig. 14 is heat treated at 1700 degrees Fahr. complete recrystallization occurs and the grains are distributed at random about the principal direction. When the strip which has been given the proper intermediate heat treatment is further cold-rolled to gage somewhere between 0.010 and 0.014 inch and heat treated, a completely random grain stucture which is characterized by a high magnetic moment providing the proper distribution of heat treatment and cold rolling has been properly followed. To explain this phenomena on the basis of preferred orientation[10] one surely would expect to find a very strongly developed fiber structure when such a specimen is X-rayed normal to its surface or when this specimen is tilted slightly with respect to the X-ray beam. The specimens for X-ray examination were always moved to and fro during exposure and some were also set at various angles to the X-ray beam. Exposures as long as a month have been made and the X-ray patterns always indicate a substantially random grain structure.

When I say that the grains in this material are fine or small I refer only to the grain structure as found in silicon steel. Some manufacturers produce sheets in which the grains are ¼ inch in diameter. I have reserved the term "fine grain" in the case of silicon steel to refer to grains in which the diameter is approximately 2 millimeters or less. From what Dr. Ruder has said in this discussion we would be led to believe that the results obtained in this investigation are distinctly characteristic of so-called controlled cold rolling as proposed by Smith, Garnett and Randall. I have followed their procedure with the results that the material was always characterized by a very low magnetic moment of the same magnitude as found in the regular hot-rolled sheets as made today. The results which I have obtained are not due to cold rolling alone but are only obtained when hot strip having the proper structure for my method is used and which is properly processed so that the final product is characterized by a magnetic moment approaching that of a single crystal; in fact, the success of this process actually depends on the proper distribution of the hot rolling, cold rolling and heat treatment rather than only emphasizing

[10]W. E. Ruder, "Magnetization and Crystal Orientation," TRANSACTIONS, American Society for Steel Treating, Vol. 8, 1925, p. 23.

cold rolling and the proper heat treatment. Hot-rolled sheets having a watt loss of 0.45 watts per pound are at the present time obtained by special selection from a whole batch and may only represent 10 to 20 per cent of the whole heat and in some cases the silicon content is well over 4.5 per cent and such material is noted for its brittleness and which makes it very difficult to handle in processing. By this new process about 70 per cent of all the coils having a watt loss well under 0.45 and the remainder under 0.50. The permeability of present day commercial sheets or cold-rolled strip at 16,000-B never exceeds 400 whereas in the new product permeabilities as high as 2700 have been obtained which is as good as some of the best silicon steel single crystals. I have also found a marked decrease in hysteresis loss as the material becomes thinner provided single crystal characteristics are present or at least the magnetic moment is large in comparison to present day materials and the grain size small. This I believe is absolutely new.

As Dr. Ruder points out the eddy currents are lowered by decreasing the thickness of the sheets but additional decline in eddy current loss is also due to small grain size. My results show that the low watt losses are associated with a high magnetic moment along the paths of easiest induction and provided the grains are not too large and the X-ray always shows a grain structure distributed entirely at random in the plane of rolling. This is a new discovery. I have found by experiment that high magnetic moment is not related to directional properties introduced by cold rolling as Dr. Ruder believes but is developed by proper distribution of heat treatment, cold rolling and hot rolling. This is only possible when the material is produced under controlled conditions so that the resulting magnetic moment of the material is high in comparison to the present day materials and in some cases approaches that of a single crystal. I have also found that the magnetic moment and the percentage of cold work are not related. At present I do not attempt to give theoretical explanations of these facts. Several interesting suggestions concerning the theoretical aspect of the problem have always been given in the direction by Professors Kunz and Williams and Mr. Ziegler. I agree with Dr. Ruder's findings that the new product has very poor properties at an angle of 45 degrees to the direction of rolling. This is as it should be since the torsion magnetization curves do not indicate a path of easiest magnetization to exist at this angle. This is illustrated in Fig. 6 of the text. I wish to point out that it is necessary to distinguish between preferred magnetic direction and preferred crystal orientation and that the former exists in this material is evident, Fig. 6. Also, a simple vector addition shows that if the crystals were equally distributed between two preferred orientations as Dr. Ruder suggests then the 45-degree direction should show much better magnetic properties. If the high magnetic moment in this material is entirely due to preferred orientation then it would be only necessary to give the hot-rolled strip a relatively large reduction and heat treated so that a strong preferred orientation would result. I have done this on many kinds of hot-rolled strip in which the material gave an area reduction ranging from 70 to 90 per cent and then heat treated at temperatures ranging from 1600 to 2000 degrees Fahr. with the result that such materials were always characterized by a very low magnetic moment. When such

a material is X-rayed in the principal direction a marked orientation is found yet the magnetic moments are very small.

I have also examined various hot-rolled sheets having especially fine magnetic properties and exceptionally poor magnetic properties and when X-rayed in the principal direction I have always found a marked preferred orientation in both kinds. The results of this study will be presented in an additional paper in the future as the results are not complete. I merely wish to point out at this time that the existence of a preferred orientation in the principal direction does not mean that a high magnetic moment will be found in the material. This means that even if preferred orientation exists in this material when X-rayed in a transverse direction the presence of orientation does not explain the phenomena of low or high magnetic moment.

Fig. 15a is an X-ray diffraction pattern made of a hot-rolled sheet which was characterized by a very low magnetic moment yet it distinctly shows a preferred orientation in which the 1 0 0 direction tends to lie in the direction of rolling. When this same material is X-rayed normal to the sheet a random orientation is the result; Fig. 15-b. When this same sheet was X-rayed in the direction parallel to the rolling direction a rather ill-defined preferred structure was also obtained. The torsion magnetization curve of this particular material is similar to the one shown in Fig. 5; that is, the magnetic moment is very small. There is another point which I wish to make. When hot-rolled strip is cold-rolled directly to gage without any heat treatments it is always characterized by a high magnetic moment; however, when it is processed according to the procedure outlined in this paper the cold-rolled material prior to the final heat treatment has a very low magnetic moment or at least much less than that of the material which is rolled directly to gage. When these two materials are X-rayed in the principal direction the fiber structure obtained is exactly the same yet one is characterized by a high magnetic moment and the other by a low magnetic moment which in some cases is very hard to detect. On the basis of preferred orientation, how can this phenomena be explained? Therefore, the presence or absence of preferred orientation does not explain this phenomena. This certainly complicates matters and proves conclusively that high magnetic moment is not associated with preferred orientation.

As Dr. Ruder points out that hot-rolled sheets have very strongly developed directional properties. I have made an extensive survey on the orientation of crystals in hot-rolled sheets and find that the grains are oriented in about the same directions in all parts of the strip when X-rayed in the principal direction. This study proves that the X-ray can detect preferred orientation when it exists it is only a question of X-raying the material in the proper directions, and certainly the X-ray is more definite and explicit in determining orientations than optical methods because numerous investigators have shown that the optical methods are at times misleading.

The work of Dr. Dean and his associates at the Bureau of Mines on powdered magnetite is very interesting and here again results seem to indicate that preferred orientation of the grains is not an essential requirement for the explanation of the effect they have found but rather their proposed explanation of "internal surface" in the way they have defined it surely helps explain this unusual phenomena.

RELATION OF CHEMICAL COMPOSITION TO THE RAIL FISSURE PROBLEM

By Earnshaw Cook

Abstract

Service data are presented indicating the influence of winter temperatures and chemical composition upon the rail fissuring effect. The development of a weighted index of chemical hardness by factoring the individual elements of the analysis is discussed. The questionable virtues of a low silicon, coarse-grained, rail steel are suggested in comparison with a thoroughly deoxidized, aluminum-killed, shallow hardening forging steel practice. Suggestions are made concerning a number of important corollaries to rail steel manufacture which might be profitably included in investigations of service performance. The importance of confining comparisons of other incidents of production to heats within similar ranges of chemical hardness is emphasized.

Certain Aspects of the Rail Fissure Problem

IT has been said that scientific research is accelerated experience, yet the lack of exact knowledge regarding the metallurgical mechanics of the causes and formation of transverse fissures in steel rails raises doubt as to the velocity with which science has contributed to an understanding of this subject. Increasing use of electrical methods for the detection of defective rails in actual service, together with the realization of high speed transportation with specially constructed, stream-lined trains, has given additional impetus to the desire of both producers and users for the solution and control of the problem.

While the numbers of rails taken from track with transverse fissures is a very small percentage of the total actually in service, this is less than what might otherwise be the case, since general practice requires that with the detection of five fissured rails in any one melt of steel, it must be removed from the main line in its entirety. This customary procedure of condemning heats upon individual performance has resulted in the removal of from 4 to 14 per cent of the total heats in the case of one important eastern railroad.

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Despite the innumerable variables of manufacture, service and inspection, a few definite conditions have been accepted as directly associated with transverse fissures. It has been shown, for example, that rails from certain mills and certain heats are more susceptible to fissure formation; that heavier wheel loads and higher speed traffic contribute to their development; and that fissures, or at least their discovery, are associated with winter temperatures.

The cold weather effects are illustrated in Table I. It should be noted that the inversion of seasonal results with the Sperry car in 1933 was caused by the inspection of considerable mileages of branch line which had not been previously included in the summaries. The routine use of electrical methods of inspection is an important adjunct to the early removal of failed rails and safer transportation.

Table I

Report of Per Cent Fissured Rails by Months for Complete System R. R. "C" Section: 105 to 127-pound

	1932				1933					
Summer Months	Without Car		With Car		Without Car		With Car		Total	
	No.	Per Cent	No.	Per Cent	No.	Per Cent	No.	Per Cent	No.	Per Cent
April	21	5.4	30	10.6	11	2.8	21	5.5	83	5.7
May	21	5.4	23	8.1	11	2.8	9	2.4	64	4.4
June	12	3.1	35	12.3	25	6.3	38	9.9	110	7.6
July	12	3.1	12	4.2	20	5.1	48	12.6	92	6.4
August	17	4.4	9	3.2	18	4.6	64	16.8	108	7.5
Sept.	34	8.7	15	5.3	21	5.3	65	16.9	135	9.3
Total Per Cent		30.2		43.6		27.0		64.3		40.9
Winter Months										
October	37	9.5	37	13.6	51	12.9	28	7.4	153	10.6
Nov.	75	19.3	10	3.5	62	15.7	32	8.4	179	12.3
Dec.	58	14.9	23	8.1	76	19.5	35	9.4	192	13.4
Jan.	25	6.4	65	22.8	34	8.6	16	4.2	140	9.7
Feb.	41	10.6	10	3.5	42	10.7	18	4.7	111	7.6
March	36	9.3	15	5.3	22	5.6	6	1.6	79	5.4
Total Per Cent		69.8		56.4		73.0		35.7		59.1

Until very recently the carrier companies have insisted upon higher chemical specifications to obtain greater girder strengths and reduced wheel erosion which have progressively lessened the impact properties of the steel while probably increasing its inherent susceptibility to fissure formation in both manufacture and service. Although the steel producers have yet made little progress in generally improved metallurgical practices to process this material, there seems to be little justification for penalizing the whole track system and the mills with high analyses to obtain better wearing properties on the curves, which are but a small percentage of the track mileage.

It is usual experience that metallurgical defects are the result of combinations of unfavorable variables uniting toward a common purpose; the individual effects cannot be determined until each has

been severally eliminated or controlled. Consequently, most mill practices are a series of compromises between opposing and conflicting conditions such as homogeneity and product yield, hardness and ductility, control of piping and surface quality, and so on.

Occasionally, some one condition overshadows all others and obstructs further progress until definite control has been accomplished. This circumstance appears to have been exemplified in a series of some 140 heats of 130-pound rails processed in 1929 for one of the southern railroads; transverse fissures had evidently been developed at such a rate that a considerable number of failures appeared before it was possible to remove all the rails from service.

A definite relation having been previously observed between drop test failures and increasing "Chemical hardness," as calculated from a weighted index of the individual elements, it was decided to develop similar comparisons concerning the prevalence of fissures in comparable groups of heats. It was assumed that, if the increasing hardness of the steel were responsible for a tendency toward either the formation of incipient transverse fissures during manufacture or the inherent lack of shock resisting properties in the material, a parallel occurrence of fissures should prevail.

"Hardness numbers" for varying chemical analyses may be calculated from the usual tensile strength formulae or from a series of previously developed factors:

T. S. = 36500 + 800 (C) + 150 (Mn) + (60 + 3.6 C) (P) + 50 (S) + 120 (Si)

Element	Hardness From Formula	Factors As Used in Calculations	
Carbon	32	0.80% x 30	= 2400
Manganese	6	0.70 x 8	= 560
Phosphorus	14	0.020 x 4	= 80
Sulphur	2	0.030 x 1	= 30
Silicon	5	0.20 x 5	= 100
		Hardness No.	= 3170

In each case the points of each element present, as usually reported by the laboratory, are multiplied by the corresponding factor. In any particular range of specifications, the variations of the physical properties follow the hardness numbers in faithful agreement, but, as might be expected, steels of widely differing composition cannot be so compared. The complete applicability of the formula remains to be proven, but its use with available data indicates that it should give results approaching the truth.

Before presenting service data, it is well to consider the appli-

cation of the "hardness factors" to the standard chemical specifications for 121-pound to 140-pound rails:

Element	1932 Chemical Specifications: Low Per Cent	Mean Per Cent	130-pound Rails High Per Cent
Carbon	0.72	0.81	0.89
Manganese	0.50	0.70	0.90
Phosphorus	0.015	0.020	0.025
Sulphur	0.020	0.030	0.040
Silicon	0.15	0.20	0.25
"Hardness Number"	2715	3200	3655
Per Cent Increase in Hardness	0	18	36

	1933 Chemical Specifications: Per Cent	Per Cent	130-pound Rails Per Cent
Carbon	0.69	0.76	0.82
Manganese	0.70	0.85	1.00
Phosphorus	0.015	0.020	0.025
Sulphur	0.020	0.030	0.040
Silicon	0.10	0.17	0.23
"Hardness Number"	2760	3155	3515
Per Cent Increase in Hardness	0	14	27

It is especially interesting to observe that the apparent concession in carbon is practically offset by the increment of manganese. The lowest possible analysis is actually increased while the important mean specification is lowered less than one and one-half per cent. As produced, a large proportion of the heats will still be definitely in the dangerous hypereutectoid range because of the transpositional effect of higher manganese contents upon the carbon eutectoid point.

The surprising differences in physical properties of rail steel from the upper and lower brackets of the specifications are clearly demonstrated by the following examples:

Heat No.	Treatment	C	Mn	P	S	Si	Hard. Nu.	Rockwell	Brinell
A	As-Rolled	0.88	0.89	0.024	0.032	0.26	"3610"	C 30	285
	Furnace cooled 1600 Degrees Fahr.							C 26	262
B	As-Rolled	0.69	0.72	0.020	0.020	0.26	"2876"	C 23	241
	Furnace cooled 1600 Degrees Fahr.							C 17	212

Heat "B" is softer and more ductile as-rolled than heat "A" after furnace cooling from 1600 degrees Fahr. Similar relationships appear between hardness numbers and tensile strengths in the as-rolled condition:

Heat Number	Hardness Number	Tensile Strength in pounds per square inch
1	2878	128,000
2	2916	128,500
3	2947	129,300
4	2957	129,000
5	2992	131,000
6	3010	136,000
7	3012	133,000
8	3059	133,000
9	3095	137,300
10	3179	138,800
11	3257	140,000

Application of the hardness factors to the first collection of field data, for which only the failed heats were available, gave an encouraging tabulation of results: (Table II.)

Table II
Railroad A: 130-Pound Rails: 1928-1929 Rollings

"Hardness No."	No. Heats	Total Rails	Per Cent Failures Of Heats	Per Cent Failures Of Rails
3000	68	..	1.5	..
3050-3350	293	..	5.2	..
3400-3700	83	..	12.0	..
Totals	444		5.9	

The second group of heats was accompanied by more detailed information, and was separated therefore into a larger number of sub-divisions. (Table III.)

Table III
Railroad B: 130-Pound Rails: 1929 Rolling

"Hardness No."	No. Heats	Total Rails	Per Cent Failures Of Heats	Per Cent Failures Of Rails
2850	4	600	0.0	0.0
2950	11	1650	36.0	0.36
3050	28	4200	36.0	0.36
3150	29	4350	48.0	1.90
3250	25	3750	56.0	2.40
3350	20	3000	67.0	3.40
3450	10	1550	90.0	6.50
3550	3	450	66.0	5.10
Total	170	..	39.0	

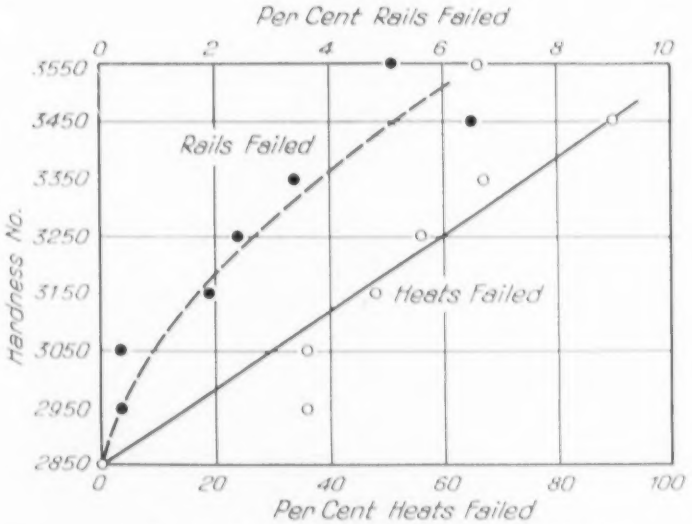

Fig. 1—Curves Illustrating Data of Table III.

No explanation can be offered for the tremendous differences in the percentages of heats removed from service by the two railroads:

the quality of the steel, the severity of the service, the condition of the road beds, the methods of inspections,—each or all may have affected the results. Nevertheless, it appears definitely established that, even without necessarily comparable conditions for each group, an increasing susceptibility for the formation of transverse fissures is associated with heats in the higher ranges of chemical specifications. The regularly mounting increment of failures per heat is dependable evidence of the influence of harder steel upon the fissuring effect. It seems especially significant that none of the four heats of lowest hardness numbers developed fissures, while eleven of the thirteen heats in the highest groups exhibited marked individual sensitivity to the defect and had to be removed from the track. The unfortunate phase of the situation for the rail mill is that, for every heat shipped on any one rolling *below* the mean carbon, the specifications require shipment of another heat *above* the mid-point. Until very recently, the railroads have been adamant toward any suggestion for generally lowering the chemical specifications. While it may have been proven that higher analyses provide better wearing qualities, it might also be well to remember that fissured heats prematurely removed from service can never be worn out in main line track.

Serious attention should likewise be devoted to the change of the A.R.E.A. silicon specifications from "0.15 per cent minimum" to the limits, "0.10 to 0.23 per cent," particularly where deoxidation with aluminum has usually been discouraged. Perhaps the following tabulations (Tables IV and V) obtained from another of the eastern roads influenced this decision: they represent summaries of the percentages of fissured heats embracing some hundreds of melts against the usual elements of the analyses, and include steel supplied from three different rail mills:

Table IV

Per Cent Fissured Heats: 130 Pound Rails: By Analyses: 1926

Carbon Per Cent	Failures Per Cent	Manganese Per Cent	Failures Per Cent	Silicon Per Cent	Failures Per Cent
.70 Max.	5.0	0.50-0.55	9.0	0.15-0.16	4.0
.71-.75	5.0	0.56-0.60	9.0	0.17-0.18	4.0
.76-.80	7.0	0.61-0.65	7.0	0.19-0.20	4.0
.81-.85	10.0	0.66-0.70	4.0	0.21-0.22	7.0
.85+	7.0	0.71-0.75	6.0	0.23-0.24	11.0
		0.76-0.80	5.0	0.25-0.26	15.0
		0.81-0.85	9.0	0.27-0.28	22.0
		0.86+	7.0	0.29-0.30	14.0
				0.31+	25.0

(Note: These figures are approximate as taken from a series of small scale graphs.)

This appears at first sight a serious indictment of even moderate

silicon in steel rails until the then standard methods of manufacture are carefully considered: these records were compiled without regard to the variation of other elements. Where for given specifications closely equivalent weights of hot metal and ferro-alloys are added to furnace and ladle, the briefest reflection upon the practices will suggest that the alloy yields and analyses will vary inversely with the relative oxidations of bath and slag as illustrated below for 353 rail heats tapped in this manner by one plant in the last six months of 1929:

Table V
Summary of Rail Heats by Silicon Ranges

No. Heats	Silicon Per Cent	Carbon Per Cent	Manganese Per Cent	Phosphorus Per Cent	Sulphur Per Cent	Hardness No.
37	0.20	0.77	0.67	0.026	0.030	3080
47	0.22	0.76	0.70	*	*	3084
49	0.24	0.78	0.71	*	*	3162
68	0.26	0.77	0.73	*	*	3158
63	0.28	0.77	0.73	*	*	3176
45	0.30	0.78	0.76	*	*	3232
44	0.32	0.78	0.75	*	*	3230

It is at once apparent that any separation of recarburized rail heats according to ladle silicons likewise segregates those melts of generally higher analyses: the lower the slag oxidation, or the higher the recarb silicon and, or, the preliminary carbon, the higher the final alloy yields become. The interesting curve of failed heats against ladle silicons is also very probably a history of the effects of generally increasing hardening elements with an aggravated susceptibility to the probable ill effects of comparatively rapid cooling on the hot beds. In answer to the objection that 200 points increase in hardness number should not necessarily account for an increase of fissured heats from 4 to 25 per cent an examination of the earlier tabulation of heats from Railroad B shows an increment of 36 per cent failed heats for the same rise in hardness.

The only really comparable records which can be advanced in reference to the debatable silicon effect must include only heats of similar hardness numbers as obtained from the data cited from Railroad B:

Table VI
Comparable Range

Ladle Silicon, Per Cent	(0.20)	0.22	0.24	0.26	0.28	(0.30)
No. Heat	(19)	19	22	25	27	(10)
Avg. "Hardness No."	(3070)	3110	3170	3190	3190	(3320)
Per Cent Heats Failed	(48)	42	50	48	52	(70)
Per Cent Rails Failed	(0.9)	2.9	2.5	2.3	2.5	(4.4)

From the preceding tabulation (Table VI), none of the silicon effects may yet justly be said to exert malignant influences upon the fissuring properties when present in reasonable percentages where hardness numbers are comparable.

Other things being equal, it is well demonstrated in plate mill practices, for example, that freedom from ingot cracks is definitely associated with the less coarsely crystalline primary structures induced by decreasing the silicon content of the steel from 0.30 per cent toward 0.12 per cent at the prohibitive expense of nonhomogeneity and segregation. A similar effect is observable for rail steel compositions, but, once the ingot has been broken down, it is debatable as to whether the phenomenon persists to increase the susceptibility of the material to fissure formation.

In the recent American Society for Metals Symposium on Grain-size, Scott[1] has reported that, other things being equal, increasing additions of silicon to high carbon steel impart a tendency toward the grain refinement lending itself to improved impact properties after hardening. While large inherent grain-size and susceptibility to deeper hardening have long been associated with silicon-killed steels in the absence of colloidal dispersions of oxides or undissolved carbides, it does not necessarily follow that such an effect is progressive with higher silicon contents.

A typical cross section of a low silicon-killed ingot of rail composition appears in the accompanying illustration. Every steel plant that has attempted to produce the so-called coarse-grained, normal steel for heat treating or carburizing is only too familiar with the excessive chipping costs and carbon segregation characteristic of aluminum-free, low silicon-killed heats. Before embarking upon a campaign for producing thousands of tons of rail ingots of this generally unsound type, it would seem most important to consider the possibilities of carefully manufacturing a thoroughly deoxidized steel for the purpose.

Here, as in most cases, the service requirements cannot be duplicated in the laboratory, but must be developed in test track under conditions of actual and comparable operation. Neither the manufacturers nor the railroads are yet in a position to render a decision as to the chemical specifications or plant practices best designed for achieving the desired practical results. The low silicon experiment, for instance, was more or less usual practice in the mills ten years

[1]Howard Scott, "Factors Determining Impact Resistance of Hardened Carbon Steels," TRANSACTIONS, American Society for Metals, Vol. 22, 1934, p. 1142.

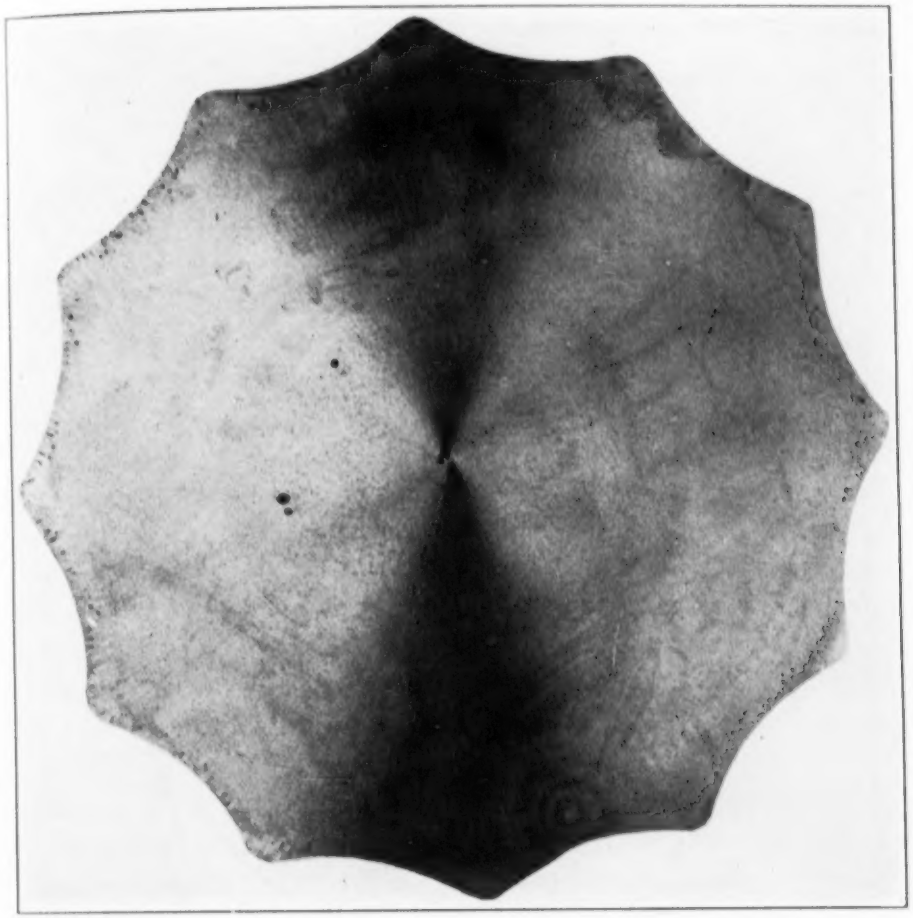

Fig. 2—Photograph of Polished and Unetched Disk from 21-Inch Round Corrugated Ingot of Low Silicon, Semi-killed Ingot Showing Defects. Ladle Analysis—Carbon 0.72, Manganese 0.76, Phosphorus 0.023, Sulphur 0.034 and Silicon 0.16.

ago, and was abandoned for the express purpose of controlling carbon segregation. Large scale, cyclical changes are frequently not comparable and usually undemonstrative. Excepting for special and costly test track installations, the performance quality of a particular heat can seldom be obtained in less than three years. In view of this situation, it would appear advantageous that the various co-operative programs supported by producers and carriers attempt to establish, as well as may be by carefully arranged experiments, a number of important corollaries to rail steel manufacture.

Such investigations might profitably include the following problems:

(1) Development of the correct chemical specifications, based on a weighted hardness index, to determine the best compromise be-

tween wearing qualities and the fissuring effect. The continued use of a brittle, hypereutectoid steel, with Izod impact properties at times less than one foot-pound in the as-rolled condition, seems ill-advised. Distinction in specifications might well be made between rails for curves and straight line track if abrasion properties require it.

(2) With the supposition that all hardening elements be held to a minimum, definite comparisons should be obtained between the present semi-killed, low silicon open-hearth ingots and a thoroughly deoxidized steel approaching forging quality. The work of numerous investigators has demonstrated the deep hardening properties of coarse-grained, silicon-killed steels. It may be that the shallow hardening, fine-grained aluminum-killed steels may prove more resistant to both thermal strains and wheel impacts. Certainly the carbon transformations will be favorably accelerated in rapid cooling on the present hot beds.

(3) The virtues of inverted, hot-top ingots should be thoroughly examined in comparison with those of standard practice. The accompanying elimination of X-rails, reduction of segregation, seconds and bottom crop losses will go far toward balancing the apparent additional cost of production.

(4) The best combination of tensile strength and elongation for steel in the as-rolled condition is usually associated with a refined grain-size for any standard operation. This will vary with the deoxidation practice in the open-hearth the total reduction from the ingot and the finishing temperature on the mills. An example of the influence of rolling mill practices upon the physical properties of eutectoid steel is shown below:

Average Finishing Temperature Degrees Fahr.	Tensile Strength Pounds per Square Inch	Per Cent Elongation in 2 Inches	Per Cent Reduction Area	Brinell Hardness
1920	128,000	6.5	9.7	255
1740	133,700	11.5	28.3	255
1510	134,000	14.0	35.7	255

With a critical temperature of approximately 1350 degrees Fahr., 130-pound rail sections are finished at 1850 to 2000 degrees Fahr. in standard mill operations,—nearly 500 degrees above the Ac point. The reasons advanced for this are the limited capacity and wear of the rolls, and the temperature differentials between the head and the flanges. With the necessary precautions, however, these objections could be sufficiently circumvented by so holding the rails to the minimum practical temperatures before the last few mill passes as to

materially refine the grains, thereby improving resistance to thermal stresses, impact and probably abrasion. The possibilities could be readily determined by service tests, which should not be discouraged by previous, spasmodic, unsuccessful attempts.

(5) Both metallurgical reasoning and the results thus far obtained verify the wisdom of expediting the tendencies of modern mill practice toward retarded cooling or normalizing from above 1000 degrees Fahr. A report of H. H. Munro[2], referring to the Sperry car inspection of French railroad systems, states: "It is significant that not a single defect in a heat treated rail has been found so far." This evidently refers to the oven process for slow cooling on the hot beds. Permitting a susceptible, coarse-grained, eutectoid steel of asymmetrical section to cool rapidly and irregularly through both the critical, secondary and blue brittleness ranges appears calculated to induce a chilled, highly strained internal state favorable to the formation of incipient thermal cracks. This condition certainly existed to a marked extent in medium manganese rails where few of numerous etch tests from the head failed to reveal frequent longitudinal and occasional transverse thermal cracks which experiment proved could be visibly eliminated, at least, by slow cooling. It is not impossible that a common macroscopic effect in medium manganese rails may be of microscopic occurrence for the slightly less susceptible high carbon steels to act as a nucleus for subsequent fatigue failure.[3]

Single refinements of practice can seldom be advocated as a panacea for metallurgical difficulties, but it would appear sensible that downward revision and control of the chemical specifications together with retarded cooling on the rolling mill hot beds could be depended upon to ameliorate the rail fissure problem. It seems evident that the high analysis effect is particularly important, and that no logical comparisons of other incidents of manufacture may be properly made excepting for heats within similar ranges of hardness as indicated by a weighted index of their alloying elements.

[2]H. H. Munro, *Railway Engineer*, September, 1933.

[3]E. S. Davenport, E. L. Roff and E. C. Bain, "Microscopic Cracks in Hardened Steel. Their Effects and Elimination," TRANSACTIONS, American Society for Metals, Vol. 22, 1934, p. 289.

MAGNETIC PROPERTIES OF IRON AS AFFECTED BY CARBON, OXYGEN AND GRAIN-SIZE

By T. D. Yensen and N. A. Ziegler

Abstract

Ring samples of iron low in sulphur and phosphorus (less than 0.003 per cent), and with varying quantities of oxygen and carbon prepared from ingots made in a high frequency induction vacuum furnace, were carefully analyzed for carbon, oxygen and grain-size after final annealing and magnetic testing. After eliminating the effect of grain-size (Figs. 1, 2, 3) the effect of oxygen and carbon on the magnetic properties is determined (Figs. 7 and 8) and developed into a ternary diagram, (Fig. 9) by means of which the net effect of grain-size is obtained (Fig. 10). The results are tied up with the previous paper (1924) on the subject by means of Fig. 13.

IN 1924 one of the authors published a paper on the magnetic properties of the ternary alloys: iron-silicon-carbon,[1] giving the first quantitative results on the effect of carbon, sulphur, phosphorus, manganese and grain-size on the magnetic properties of these alloys. While it was stated in the paper that the results should be regarded as a first approximation only, it is perhaps indicative of the difficulties involved in this problem that no new curves have been published since that time, and that the original curves can be found in many recent publications, including the International Critical Tables, The Alloys of Iron Monograph on Alloys of Iron and Silicon by Greiner and Stoughton, and in Messkin-Kussmann's Ferromagnetische Legierungen. Nevertheless it was strongly felt that the curves needed confirmation and revision on the basis of more reliable data obtainable by means of improved apparatus and technique and that there was urgent need for including the effect of oxygen. In the period 1925-27 the co-author in co-operation with P. H. Brace developed an

[1]T. D. Yensen, "The Magnetic Properties of the Ternary Alloys of Iron, Silicon and Carbon," *Transactions*, American Institute of Electrical Engineers, Vol. 43, 1924, p. 145.

A paper presented before the Sixteenth Annual Convention of the Society held in New York City, October 1 to 5, 1934. Of the authors who are members of the society, T. D. Yensen is manager of the magnetic division and N. A. Ziegler until recently was research metallurgist for the Westinghouse Electric and Manufacturing Co., East Pittsburgh, Pa. Manuscript received August 4, 1934.

improved high frequency type of induction vacuum furnace (the "Bell Jar" furnace) in which 7 kilograms of metal could be melted under a pressure of 0.001 millimeter of mercury.[2] They also developed a method for analyzing gases given off by metals when heated and melted in vacuum. The latter was later developed into a vacuum fusion method for oxygen determination[3] having an accuracy of ±0.001 per cent, that also served for hydrogen and nitrogen determination. These improved apparatus, together with the previously described "Yensen" method of carbon determination, further improved upon by one of the authors,[4] provided the means for carrying out the above purpose.

The present report is devoted to unalloyed iron only, and will be followed in due time by reports on iron-silicon alloys. With a few exceptions the samples were prepared by melting electrolytic iron together with the necessary addition agents either in vacuum or in hydrogen and the ingots were forged to bars from which ring samples 1 inch outside diameter by ¾ inch inside diameter by 1 inch high were machined.

In preparing and heat treating iron samples, a number of factors enter that are difficult to control. For example, although we can analyze for carbon to within ±0.0005 per cent and for oxygen to within ±0.001 per cent we cannot deliberately control the actual composition within these limits. For this reason, the method in this case, as in the earlier investigation, was to prepare alloys as close to the desired composition as possible, subject them to different heat treatments, and to obtain magnetic, chemical and metallographic data on every sample *in the final state* so as to be sure to have composition and grain-size correspond to the magnetic test results.

The heat treatment consisted in annealing the samples either in vacuum at 900 to 1100 degrees Cent. (1650-2010 degrees Fahr.) or in hydrogen at 900 to 1400 degrees Cent. (1650-2550 degrees Fahr.) and for various lengths of time followed by slow cooling in order to get the particular magnetic properties desired. The principle followed is that the magnetic properties are functions of the composition and structure, so that, if these can be accurately determined, reliable

[2]P. H. Brace and N. A. Ziegler, "Applications of a High Vacuum Induction Furnace to the Study of Gases and Metals," *Proceedings*, Institute of Metals Division of American Institute of Mining and Metallurgical Engineers, 1928, p. 544.

[3]N. A. Ziegler, "Improved Methods for the Analysis of Gaseous Elements in Metals," *Transactions*, American Electrochemical Society, Vol. 62, September 1932, p. 109.

[4]N. A. Ziegler, "Recent Developments in the Analysis of Iron and Iron Alloys," *Transactions*, American Electrochemical Society, Vol. 56, 1929, p. 231.

relationships for the determination of the magnetic properties can be obtained irrespective of the particular heat treatment used.

The results for about 100 ring samples of unalloyed iron are reported in Table I, II, and III. Columns 2, 3, 5 and 6 of each table give the results of the chemical analyses; Column 4 the result of the microscopic examination, grains per square millimeter (N); and Columns 7, 8, 9, 10 and 11 the results of the magnetic tests:

$\mu_{H=10}$ = permeability for H = 10
$\mu_{max.}$ = max. permeability
$\rho_{min.}$ = min. reluctivity $\left(= \frac{1}{\mu_{max.}}\right)$
H_c = coercive force (oersteds) for B = 10,000
W_h = hysteresis loss (ergs per cubic centimeter per cycle) for B = 10,000

The sulphur content of the samples as shown by the chemical analysis of ten samples selected at random was nearly constant and in no case was it over 0.003 per cent. Phosphorus and manganese were also present in amounts less than 0.003 per cent. It should also be noted from Tables I, II and III that the hydrogen and nitrogen contents of all the samples (Columns 5 and 6) are very low and uniform. For these reasons the effect of all these elements may be regarded as small and as nearly constant. By disregarding them a constant small shift may have been introduced in the curves, but not sufficient to affect the results appreciably. Furthermore, grain or lattice orientation which chiefly affects the high flux density permeability, was not taken into account in this investigation. The samples reported here all have more or less randomly oriented grains.

This leaves the three main factors, carbon, oxygen and grain-size, whose effect on the magnetic properties are to be determined from the original data as given in Table I, II, and III. In order to do this it was necessary to select and group together samples for which two of the factors, carbon and oxygen, are nearly constant and then to plot curves for the third factor, grain-size, against the various magnetic characteristics: maximum permeability ($\mu_{max.}$), min. reluctivity ($\rho_{min.}$), coercive force (H_c) and hysteresis loss (W_h). However, as minimum reluctivity is nearly proportional to the coercive force (Fig. 4) and to hysteresis loss (Fig. 5) as long as held within this group of materials, it is sufficient to plot curves only for grain-size versus minimum reluctivity for constant carbon and oxygen, as has been done in Figs. 1, 2 and 3.

Table I

Chemical, Microscopic and Magnetic Properties of Specimens

Group No.	Composition Range	Sample No.	Per Cent Carbon	Per Cent Oxygen	Grains per Sq. mm.	Per Cent Hydrogen	Per Cent Nitrogen	Original Values (7) $\mu_{H=10}$	(8) μ_{max}	(9) $\rho_{max.} \times 10^6$	(10) H_c	(11) W_h
		(1)	(2)	(3)	(4)	(5)	(6)					
1.	C less than 0.002% O_2 less than 0.005%	AQ-1-f	0.0010	0.0032	1.52	0.0011	0	1551	132700	0.75		
		-g	0.0004	0.0047	1.45	0.0012	0	1520	121000	0.83		
		-h	0.0011	0.0024	1.14	0.0010	0	1490	131000	0.76		
		-k	0.0011	0.0028	0.76	0.0008	0	1550	175000	0.57		
		-l	0.0020	0.0048	0.76	0.0008	0	1500	177900	0.56		
		-n	0.0010	0.0022	0.63	0.0011	0	1545	207500	0.48	0.0278	70
		-o	0.0017	0.0039	1.93	0.0007	0	1421	93900	0.06		
		-t	0.0017	0.0017	2.32	0.0007	0	1515	87800	1.14		
		-v	0.0015	0.0019	2.34	0.0005	0	1490	84200	1.19		
		AQ-1-J	0.0012	0.0046	3.22	0.0007	0	1425	70000	1.43		
		AQ-158-J	0.0004	0.0029	2.57	0.0008	0	1520	80500	1.24	0.032	115
		AQ-159-L	0.0009	0.0045	2.10	0.0007	0	1400	104000	0.96	0.044	133
		Average	0.0012	0.0031	1.73–1.77					0.91	0.035	106
2.	C less than 0.002% O_2 0.005 to 0.010%	AQ-1-d	0.0019	0.0075	2.08	0.0008	0	1550	79200	1.26		
		-e	0.0006	0.0095	2.59	0.0012	0	1491	79500	1.26		
		-i	0.0014	0.0072	1.26	0.0008	0	1500	146600	0.68		
		-j	0.0009	0.0053	1.01	0.0012	0	1555	160500	0.62		
		-r	0.0018	0.0035	2.34	0.0007	0	1465	86000	1.16		
		-w	0.0007	0.0070	2.91	0.0008	0	1555	85600	1.20		
		-x	0.0019	0.0063	2.54	0.0007	0	1534	76600	1.60		
		AQ-1-G	0.0014	0.0063	3.53	0.0008	0	1415	62600	1.60		
		-H	0.0012	0.0099	6.88	0.0008	0	1413	34700	2.88		
		-I	0.0019	0.0063	4.88	0.0011	0	1450	49600	2.02		
		AQ-158-D	0.0011	0.0060	3.00	0.0005	0	1540	74000	1.35		
		-F	0.0017	0.0054	2.86	0.0006	0	1530	79200	1.26		
		-I	0.0012	0.0078	3.05	0.0009	0	1542	72500	1.38		
		AQ-159-F	0.0020	0.0087	1.01	0.0003	0	1460	146800	0.68	0.038	112
		Average	0.0013	0.0072	2.88–1.01					1.34	0.038	112
3.	C less than 0.002% O_2 0.010 to 0.015%	AQ-1-c	0.0016	0.0137	3.66	0.0008	0	1523	48900	2.04		
		AQ-1-C	0.0012	0.0124	3.56	0.0007	0	1465	51000	1.96		
		AQ-158-B	0.0013	0.0130	2.26	0.0005	0	1540	63000	1.34	0.062	192
		-E	0.0018	0.0138	2.11	0.0007	0	1561	65500	1.53		
		AQ-158-H	0.0012	0.0148	3.19	0.0002	0	1470	64200	1.56		
		AQ-159-A	0.0020	0.0102	1.39	0.0011	0	1435	85000	1.18	0.031	172
		-B	0.0019	0.0107	1.26	0.0007	0	1400	87100	1.15	0.051	172
		I-8-A	0.0016	0.0149	3.42	0.0009	0.0005	1448	46800	2.13	0.120	407
		-D	0.0015	0.0148	3.39	0	0	1490	49000	2.04	0.120	412
		Average	0.0016	0.0132	2.69–2.34					1.66	0.081	271

Table I (Continued)
Chemical, Microscopic and Magnetic Properties of Specimens

Group No.	Composition Range	Sample No. (1)	Per Cent Carbon (2)	Per Cent Oxygen (3)	Grains per Sq. mm. (4)	Per Cent Hydrogen (5)	Per Cent Nitrogen (6)	Original Values $\mu_{H=10}$ (7)	Original Values $\mu_{max.}$ (8)	Original Values $\rho_{min.} \times 10^5$ (9)	Original Values H_c (10)	Original Values W_h (11)
4.	C less than 0.0020%	AQ-1-B	0.0019	0.0169	4.23	0.0006	0	1465	37200	2.69		
		AQ-159-H	0.0019	0.0181	1.09	0.0009	0	1425	63700	1.57	0.059	187
	O_2 0.015 to 0.025%	AQ-159-J	0.0013	0.0208	0.23	0.0008	0	1455	76400	1.31	0.051	159
		I-8-B	0.0018	0.0161	3.00	0	0	1439	46700	2.14	0.122	430
		I-8-C	0.0018	0.0162	3.43	0.0004	0	1400	41700	2.40	0.105	391
		Average	0.0017	0.0177	2.40–1.94					2.02	0.083	292
5.	C less than 0.0020%	I-5-D	0.0011	0.0413	4.44	0.0004	0	1457	38100	2.62	0.126	445
		I-143-A	0.0007	0.0439	20.81	0.0007	0	1506	12100	8.27	0.239	755
	O_2 0.040 to 0.045%	I-144-A	0.0009	0.0450	14.23	0.0005	0	1667	16700	5.99	0.331	971
		I-144-B	0.0010	0.0443	14.68	0.0002	0	1472	16100	6.20	0.325	1047
		Average	0.0009	0.0436	13.29					5.77	0.255	804

Table II
Chemical, Microscopic and Magnetic Properties of Specimens

Group No.	Composition Range	Sample No. (1)	Per Cent Carbon (2)	Per Cent Oxygen (3)	Grains per Sq. mm. (4)	Per Cent Hydrogen (5)	Per Cent Nitrogen (6)	Original Values μ_{10} (7)	Original Values $\mu_{max.}$ (8)	Original Values $\rho_{min.} \times 10^5$ (9)	Original Values H_c (10)	Original Values W_h (11)
1.	C 0.0020 to 0.0040%	AQ-1-a	0.0026	0.0018	1.14	0.0006	0	1430	134400	0.74		
		-b	0.0039	0.0034	1.52	0.0007	0	1416	96900	1.03		
	O_2 less than 0.005%	-m	0.0025	0.0040	3.22	0.0006	0	1446	78300	1.28		
		Average	0.003	0.003	1.96					1.02		
2.	C 0.0020 to 0.0040%	AQ-1-D	0.0024	0.0084	4.23	0.0005	0	1400	60000	1.66		
		AQ-158-A	0.0025	0.0072	4.11	0.0003	0	1520	56500	1.77	0.070	219
	O_2 0.005 to 0.010%	-C	0.0024	0.0079	4.82	0.0005	0	1420	55200	1.81		
		AQ-159-C	0.0021	0.0086	1.14	0.0002	0	1485	98000	1.02	0.044	138
		Average	0.00235	0.008	3.58–2.63					1.57		
3.	C 0.0020 to 0.0040%	AQ-1-A	0.0021	0.0183	3.00	0.0006	0	1495	37000	2.70		
		-E	0.0031	0.0119	0.75	0.0006	0	1435	65700	1.52		
	O_2 0.015 to 0.020%	-F	0.0024	0.0131	0.30	0.0012	0	1470	73500	1.36		
		AQ-159-G	0.0021	0.0190	0.18	0.0004	0	1410	65500	1.50	0.046	156
		Average	0.0026	0.0156	1.06–0.18					1.77		

	0.020%	AQ-159-G	0.0021	0.0190	0.18	0.0004	0	1410	65500	1.50	0.046	156
		Average	0.0026	0.0156	1.06–0.18					1.77		

Table II (Continued)

Chemical, Microscopic and Magnetic Properties of Specimens

Group No.	Composition Range	Sample No. (1)	Per Cent Carbon (2)	Per Cent Oxygen (3)	Grains per Sq. mm. (4)	Per Cent Hydrogen (5)	Per Cent Nitrogen (6)	Original Values $\mu_{H=10}$ (7)	Original Values $\mu_{max.}$ (8)	Original Values $\rho_{min.} \times 10^6$ (9)	Original Values H_c (10)	Original Values W_h (11)
		I-6-A	0.0022	0.0345	6.14	0.0002	0.0008	1590	21400	4.67	0.290	956
		7-A	0.0038	0.0372	5.00	0.0006	0.0009	1608	22400	4.46	0.270	853
4.	C 0.0020 to 0.0040%	9-A	0.0038	0.0218	0.63	0	0	1610	30000	3.33	0.182	572
	O_2 0.025 to 0.030%	9-B	0.0037	0.0263	0.65	0	0	1597	30600	3.26	0.176	562
		9-D	0.0022	0.0259	2.55	0	0	1606	28900	3.46	0.183	580
		25-A	0.0029	0.0244	3.98	0.0006	0	1453	23700	4.21	0.221	640
		25-C	0.0021	0.0253	0.34	0	0.0001	1490	33300	3.00	0.164	491
		F-52-A	0.0022	0.0380	2.47	0	0	1587	27400	3.64	0.165	522
		-C	0.0035	0.0267	3.58	0.0008	0.0002	1587	23500	4.25	0.176	589
		Average	0.0029	0.029	2.82					3.80	0.203	640
		I-6-C	0.0028	0.0576	8.40	0.0008	0.0003	1590	20900	4.77	0.300	970
5.	C 0.0020 to 0.0040%	I-22-B	0.0040	0.0576	21.40	0.0006	0	1497	12700	7.84	0.409	1225
	O_2 0.05 to 0.10%	F-51-A	0.0038	0.0810	15.80	0	0	1581	14700	6.76	0.361	1100
		F-51-B	0.0037	0.0751	15.80	0	0	1551	15600	6.41	0.321	1040
		Average	0.0036	0.0678	15.28					6.45	0.348	1079
		I-5-A	0.0028	0.0001	10.00	0.1198	0.0005	1526	18200	5.50	0.265	827
6.	C 0.0020 to 0.0040%	I-6-B	0.0038	0.0005	12.65	0.1158	0.0003	1520	16300	6.14	0.315	1050
	O_2 0.10% and over	I-6-D	0.0030	0.0007	7.80	0.1131	0	1588	19780	5.05	0.308	1020
		I-8-C	0.0034	0	0.53	0.1128	0	1570	28470	3.51	0.243	596
		I-143-B	0.0033	0.0002	23.12	0.1162	0	1570	12180	8.20	0.331	1042
		Average	0.0030		10.82	0.1156				5.68	0.292	1107

Table III

Chemical, Microscopic and Magnetic Properties of Specimens

Group No.	Composition Range	Sample No. (1)	Per Cent Carbon (2)	Per Cent Oxygen (3)	Grains per Sq. mm. (4)	Per Cent Hydrogen (5)	Per Cent Nitrogen (6)	Original Values $\mu_{H=10}$ (7)	Original Values $\mu_{max.}$ (8)	Original Values $\rho_{min.} \times 10^6$ (9)	Original Values H_c (10)	Original Values W_h (11)
		I-40-A	0.0073	0.0026	2.00	0	0	1555	30500	3.27	0.214	694
1.	C 0.0040 to 0.0080%	I-41-A	0.0043	0.0086	4.49	0.0005	0.0005	1500	21700	4.60	0.276	881
	O_2 less than 0.0100%	I-42-A	0.0047	0.0064	3.50	0.0007	0.0003	1545	23500	4.25	0.266	846
		I-43-A	0.0045	0.0048	4.33	0.0003	0.0004	1528	22500	4.43	0.263	844
		I-35-B	0.0048	0.0056	4.07	0	0	1556	22400	4.45	0.254	823
		Average	0.0051	0.0056	3.68					4.20	0.255	818

Table III (Continued)
Chemical, Microscopic and Magnetic Properties of Specimens

Group No.	Composition Range	Sample No.	Per Cent Carbon	Per Cent Oxygen	Grains per Sq. mm.	Per Cent Hydrogen	Per Cent Nitrogen	Original Values $\mu_{H=10}$	Original Values $\mu_{max.}$	Original Values $\rho_{min.} \times 10^6$	Original Values H_c	Original Values W_h
		(1)	(2)	(3)	(4)	(5)	(6)	(7)	(8)	(9)	(10)	(11)
2.	C 0.0040 to 0.0080%	I-39-A	0.0049	0.0140	5.95	0.0004	0.0008	1535	19900	5.01	0.278	860
		I-40-B	0.0046	0.0172	7.55	0.0009	0.0011	1549	16800	5.93	0.321	1060
	O_2 0.0100 to 0.0200%	I-41-B	0.0042	0.0187	4.28	0.0005	0.0007	1550	21500	4.63	0.291	896
		I-43-B	0.0061	0.0111	4.74	0.0003	0	1565	20800	4.78	0.288	880
		Average	0.005	0.015	5.63					5.09	0.295	924
3.	C 0.0040 to 0.0080%	I-25-B	0.0057	0.0282	2.56	0.0002	0.0003	1470	22500	4.44	0.221	640
		I-25-D	0.0043	0.0236	2.50	0	0.0006	1465	23600	4.22	0.202	599
	O_2 0.0200 to 0.0300%	I-42-B	0.0063	0.0203	4.57	0.0007	0	1559	20600	4.83	0.285	874
		F-35-A	0.0062	0.0232	4.75	0.0005	0	1498	21300	4.67	0.272	825
		F-36-A	0.0052	0.0246	2.05	0.0003	0.0008	1529	23000	4.34	0.294	907
		F-36-B	0.0053	0.0213	5.55	0.0007	0.0004	1542	20300	4.71	0.288	808
		Average	0.0055	0.0235	3.67					4.57	0.260	776
4.	C 0.0040 to 0.0080%	I-7-B	0.0053	0.0421	2.21	0.0004	0.0003	1600	21200	4.70	0.280	900
		I-7-C	0.0063	0.0456	1.83	0.0011	0.0004	1601	22200	4.49	0.265	863
	O_2 0.0300 to 0.0500%	I-7-D	0.0065	0.0526	1.92	0.0004	0	1600	21900	4.55	0.266	861
		F-51-C	0.0056	0.0388	5.93	0.009	0	1600	18000	5.50	0.318	939
		F-52-B	0.0072	0.0372	7.65	0.003	0.0002	1543	17100	5.85	0.296	768
		Average	0.0060	0.0373	3.91					5.01	0.285	866
5.	C 0.0040 to 0.0080%	I-9-C	0.0040	0.0636	0.13	0	0	1597	29300	3.41	0.183	575
		I-22-C	0.0042	0.0526	5.07	0.0004	0	1510	12600	7.88	0.403	1178
	O_2 0.050 to 0.070%	I-24-C	0.0045	0.0596	0.38	0	0	1500	24200	4.13	0.199	595
		I-24-D	0.0043	0.0521	0.26	0.0003	0.0002	1468	26300	3.80	0.180	537
		Average	0.0042	0.0556	1.46					4.80	0.282	721
6.	C 0.0040 to 0.0080%	I-21-D	0.0076	0.1192	2.75	0.0011	0	1467	16100	6.18	0.315	913
		I-22-A	0.0065	0.1384	4.85	0.0006	0.0001	1487	12800	7.79	0.409	1200
	O_2 0.09% and over	I-23-C	0.0049	0.0928	1.79	0.0006	0	1515	18700	5.32	0.277	812
		Average	0.0063	0.117	3.13					6.40	0.334	975

Table I and Fig. 1 contain the results for the first division of samples with carbon less than 0.002 per cent and oxygen increasing in five groups from less than 0.005 to 0.045 per cent.

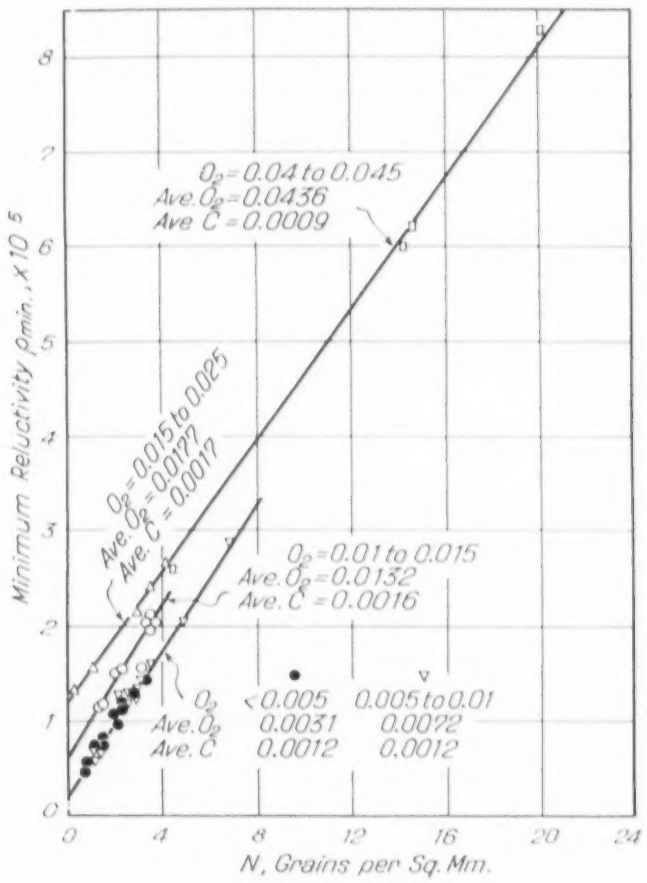

Fig. 1—$\rho_{min.}$ Versus N. C less than 0.002 Per Cent.

Table II and Fig. 2, similarly, contain the results for the second division with carbon between 0.002 and 0.004 per cent and oxygen increasing in 6 groups from less than 0.005 to about 0.15 per cent.

Table III and Fig. 3 contain the results of the third division with carbon between 0.004 and 0.008 per cent and oxygen increasing in six groups from less than 0.01 to about 0.15 per cent.

In Figs. 1, 2 and 3 by extrapolating to N = O the effect of grain-size (or grain boundaries) is eliminated and the intercepts give the net effect of the combined carbon and oxygen content for the particular group. For each division the carbon content is kept within a narrow range so that the intercepts of the curves with the ordinate

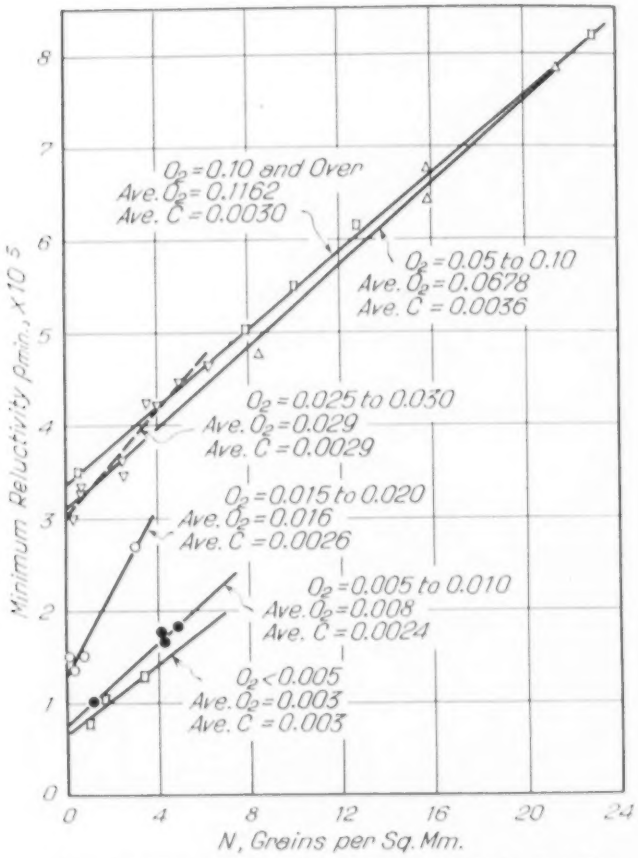

Fig. 2—$\rho_{min.}$ Versus N. C 0.002 to 0.004 Per Cent.

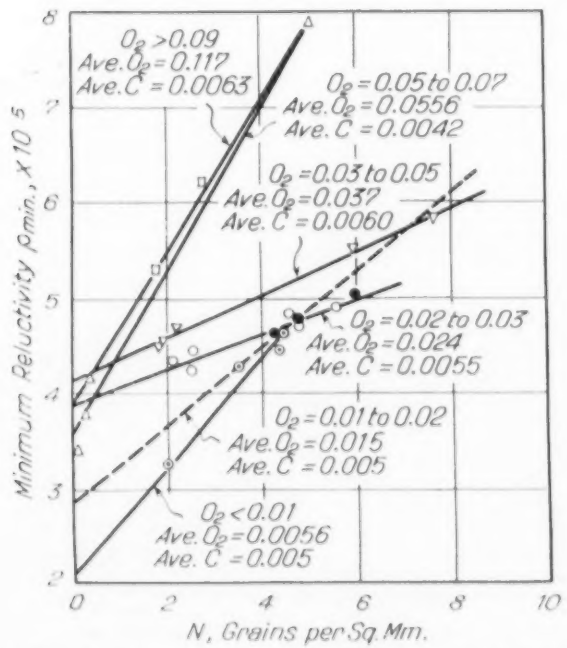

Fig. 3—$\rho_{min.}$ Versus N. C 0.004 to 0.008 Per Cent.

axis give the effect of oxygen for the following ranges of carbon content:

Fig. 1—Carbon less than 0.002 per cent; average carbon—0.0013 per cent
Fig. 2—Carbon 0.002 to 0.004 per cent; average carbon—0.0029 per cent
Fig. 3—Carbon 0.004 to 0.008 per cent; average carbon—0.0053 per cent

A summary of the results of Figs. 1, 2 and 3 is given in Table IV

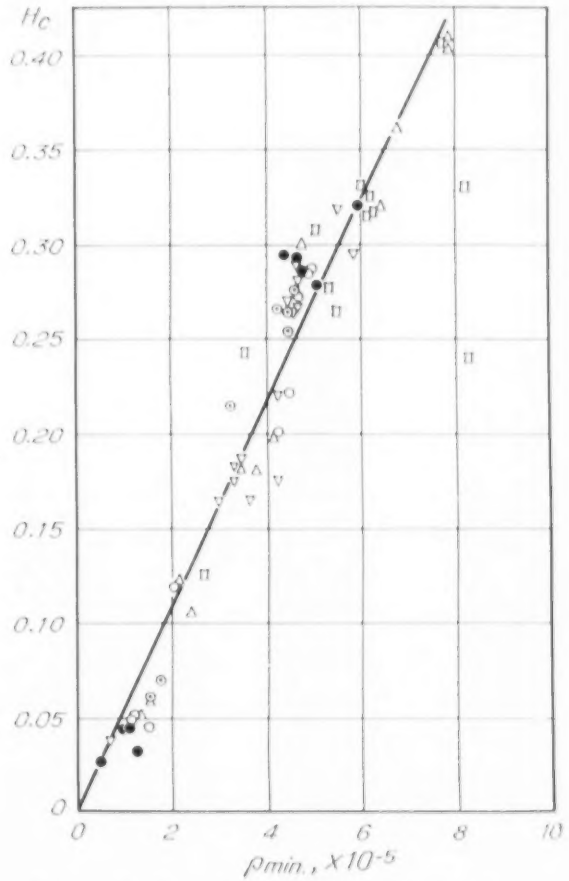

Fig. 4—Hc Versus $\rho_{min.}$ $H_c = 0.054 \times 10^5 \rho_{min.}$

where the slope is given for each individual curve together with the intercepts and the average slope for each of the three divisions.

The above intercepts have been plotted in Fig. 6 showing minimum reluctivity versus oxygen for the above three divisions of carbon content. From the intercepts of these curves with the ordinate the net effect of carbon is obtained. This has again been plotted in Fig. 7 together with the effect of carbon for various oxygen contents from 0.01 to 0.10 per cent. From this the net effect of oxygen for

Table IV

Effect of Grain-Size and Oxygen on Minimum Reluctivity ($\rho_{min.}$) for Constant Carbon Contents

Group No.	Average Carbon Per Cent	No. Pts.	Per Cent Oxygen — Range	Per Cent Oxygen — Average	$\rho_{min.}$ Versus Nitrogen (From Fig. 1)
A. Carbon less than 0.002 Per Cent					
1	0.0012	12	less than 0.005	0.0031	$\rho_{min.} = (0.38\ N + 0.25) \times 10^{-5}$
2	0.0013	15	0.005-0.010	0.0072	$\rho_{min.} = (0.38\ N + 0.25) \times 10^{-5}$
3	0.0016	9	0.010-0.015	0.0132	$\rho_{min.} = (0.40\ N + 0.60) \times 10^{-5}$
4	0.0017	5	0.015-0.025	0.0177	$\rho_{min.} = (0.35\ N + 1.20) \times 10^{-5}$
5	0.0009	4	0.04 -0.045	0.0436	$\rho_{min.} = (0.35\ N + 1.20) \times 10^{-5}$
Grand Average	0.0013	45			$\rho_{min.} = 0.37\ N \times 10^{-5}$ + effect of oxygen for Average Carbon = 0.0013%
B. Carbon 0.002 to 0.004 Per Cent					(From Fig. 2)
1	0.0030	3	less than 0.005	0.003	$\rho_{min.} = (0.19\ N + 0.65) \times 10^{-5}$
2	0.0024	4	0.005-0.010	0.008	$\rho_{min.} = (0.22\ N + 0.80) \times 10^{-5}$
3	0.0026	4	0.015-0.020	0.0156	$\rho_{min.} = (0.47\ N + 1.20) \times 10^{-5}$
4	0.0029	9	0.025-0.030	0.0290	$\rho_{min.} = (0.29\ N + 3.00) \times 10^{-5}$
5	0.0036	4	0.050-0.100	0.0680	$\rho_{min.} = (0.22\ N + 3.10) \times 10^{-5}$
6	0.0030	5	less than 0.100	0.1156	$\rho_{min.} = (0.21\ N + 3.40) \times 10^{-5}$
Grand Average	0.0029	29			$\rho_{min.} = 0.28\ N \times 10^{-5}$ + effect of oxygen for Average Carbon = 0.0029%
C. Carbon 0.004 to 0.008 Per Cent					(From Fig. 3)
1	0.0051	5	less than 0.010	0.0056	$\rho_{min.} = (0.60\ N + 2.1\ ) \times 10^{-5}$
2	0.0050	4	0.01 -0.02	0.0150	$\rho_{min.} = (0.40\ N + 2.85) \times 10^{-5}$
3	0.0055	6	0.02 -0.03	0.0235	$\rho_{min.} = (0.16\ N + 3.90) \times 10^{-5}$
4	0.0060	5	0.03 -0.05	0.0373	$\rho_{min.} = (0.22\ N + 4.15) \times 10^{-5}$
5	0.0042	4	0.05 -0.07	0.0556	$\rho_{min.} = (0.85\ N + 3.60) \times 10^{-5}$
6	0.0063	3	less than 0.09	0.1170	$\rho_{min.} = (0.78\ N + 3.90) \times 10^{-5}$
Grand Average	0.0053	27			$\rho_{min.} = 0.50\ N \times 10^{-5}$ + effect of oxygen for Average Carbon = 0.0053%

various carbon contents, from 0.001 to 0.006 per cent, has been plotted in Fig. 8, and finally the ternary diagram giving the combined net effect of carbon and oxygen is shown in Fig. 9. In the last three figures ordinates are given for coercive force and hysteresis loss in addition to minimum reluctivity, based on Figs. 4 and 5, namely:

$$H_c = 0.054 \times 10^5 \rho_{min.}$$
$$W_h = 164 \times 10^5 \rho_{min.}$$

Table V

Effect of Grain-Size on $\rho_{min.}$, H_c and W_h

Carbon Per Cent	Oxygen Per Cent	$\rho_{min.}$ From Table 4	$H_c = .054 \times 10^5\ \rho_{min.}$ (See Fig. 4) H_c Calculated	$W_h = 164 \times 10^5\ \rho_{min.}$ (See Fig. 5) W_h Calculated	
Less than 0.002	0.002 to 0.05	0.37 N × 10^{-5}	0.020 N	61 N	
0.002 to 0.004	0.002 to 0.12	0.28 N × 10^{-5}	0.015 N	46 N	
0.004 to 0.008	0.002 to 0.12	0.50 N × 10^{-5}	0.027 N	82 N	
0.004	0.04	0.383 N × 10^{-5}	0.02 N	63 N	mean

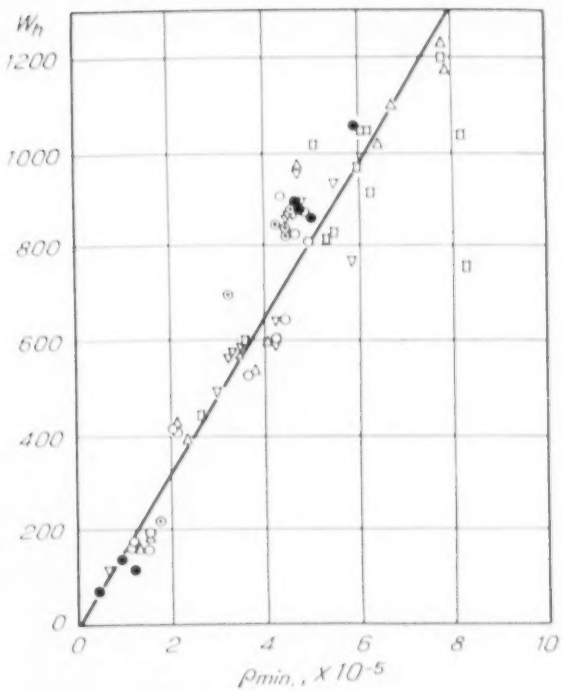

Fig. 5—Wh Versus $\rho_{min.}$ Wh = 164 × 10^5 $\rho_{min.}$

Table VI

Average Values from Original Data (From Tables 1, 2, 3)

A. Carbon Less Than 0.002 Per Cent

Group No.	No. Samples	Per Cent Carbon	Per Cent Oxygen	N Grains per Sq. mm.	$\rho_{min.}$ × 10^5	Hc	Wh	Combined Effect of C and O_2 (From Fig. 9) $\rho_{min.}$ × 10^5	Effect of Grain Size (N) (Plotted in Fig. 10) $\rho_{min.}$ × 10^5
1	12	0.0012	0.0031	1.23	0.91	.....	...	0.25	0.66
	3	0.0008	0.0032	1.77	0.89	0.035	106	....	....
2	15	0.0013	0.0072	2.88	1.34	.....	...	0.37	0.97
	1	0.0020	0.0087	1.01	0.68	0.038	112	....	....
3	9	0.0016	0.0132	2.69	1.66	.....	...	0.75	0.91
	5	0.0016	0.0127	2.34	1.61	0.081	271	....	....
4	5	0.0017	0.0177	2.40	2.02	.....	...	1.15	0.87
	4	0.0017	0.0174	1.94	1.85	0.083	292	....	....
5	4	0.0009	0.0436	13.29	5.77	0.255	804	1.17	4.60
	3	0.0009	0.0444	16.60	6.82	.....	...	1.17	5.65

B. Carbon 0.002 to 0.004 Per Cent

1	3	0.0030	0.0030	1.96	1.02	.....	...	0.80	0.22
2	4	0.0024	0.0080	3.58	1.57	.....	...	0.68	0.89
		0.0023	0.0079	2.63	1.40	0.057	179	....	....
3	4	0.0026	0.0156	1.06	1.77	.....	...	1.25	0.52
	1	0.0021	0.0190	0.18	1.50	0.046	156	....	....
4	9	0.0029	0.0290	2.82	3.80	0.203	640	2.90	0.90
5	4	0.0036	0.0678	15.28	6.45	0.348	1079	3.52	2.93
6	5	0.0030	0.1156	10.82	5.68	0.292	1107	3.35	2.33

C. Carbon 0.004 to 0.008 Per Cent

1	5	0.0051	0.0056	3.68	4.20	0.255	818	2.18	2.02
2	4	0.0050	0.0150	5.63	5.09	0.295	924	2.70	2.39
3	6	0.0055	0.0235	3.67	4.57	0.260	776	3.80	0.77
4	5	0.0060	0.0373	3.91	5.01	0.285	866	3.90	1.11
5	4	0.0042	0.0556	1.46	4.80	0.282	721	3.75	1.05
6	3	0.0063	0.117	3.13	6.40	0.334	975	4.10	2.30

From the above figures it will be seen that the effect of carbon and oxygen is not additive. In the presence of oxygen from 0.015 to 0.05 per cent the effect of carbon appears to be greater than in the absence of oxygen.

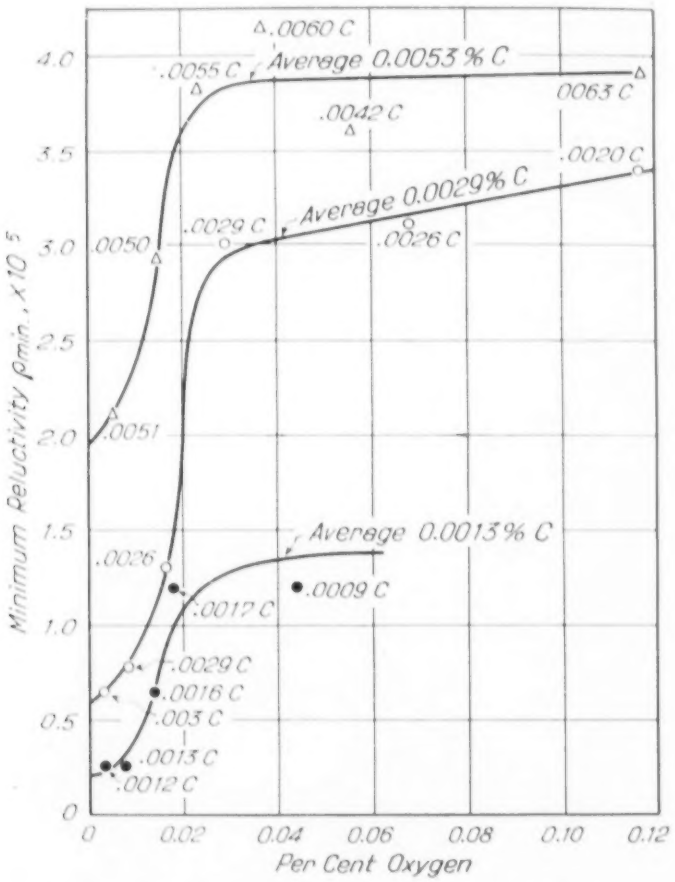

Fig. 6—Effect of O_2 on $\rho_{min.}$ for Constant Carbon.

In Table VI the average values for each group have been listed for the purpose of calculating the effect of grain-size on the basis of Fig. 9 and these values have been plotted in Fig. 10 together with the previously published curve[5] for the effect of grain-size, namely:

$$\rho_{min.} = 0.6 \times 10^{-5} \sqrt{N}$$
$$W_h = 100 \sqrt{N}$$

It would appear from the present results that the first power relation is more nearly correct than the ½ power, and that the effect depends on the carbon content.

[5]T. D. Yensen, "Does Grain Size Influence the Magnetic Properties of Iron," *Metals and Alloys*, Vol. 1, May 1930, p. 493.

Figs. 11 and 12 show respectively the net effect of carbon and oxygen on the maximum permeability. The highest actual value obtained in this series was 220,000, (hysteresis loss 70 ergs) with a sample containing 0.001 per cent carbon, 0.002 per cent oxygen and

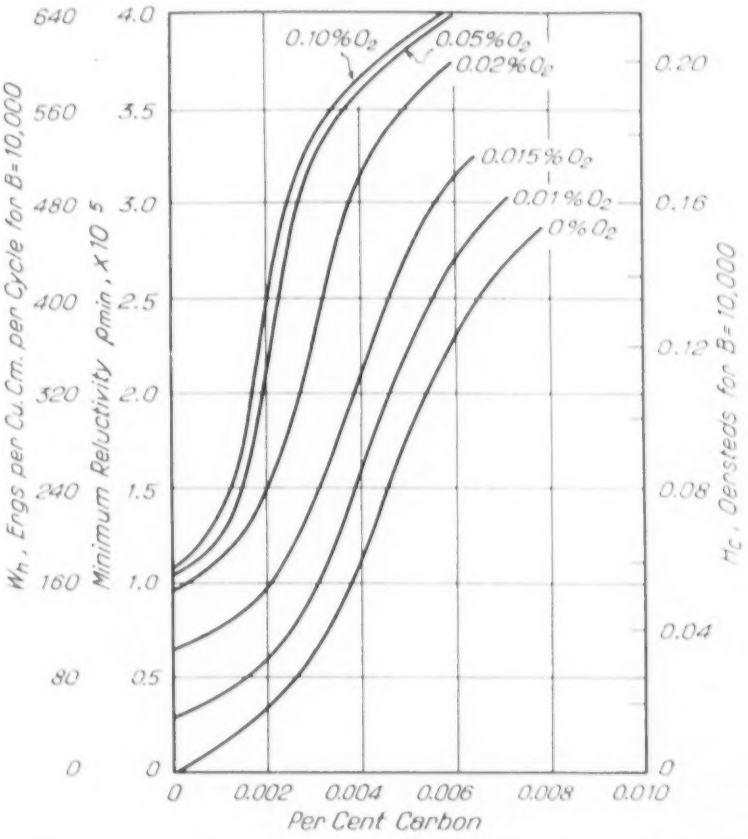

Fig. 7—Effect of Carbon on $\rho_{min.}$, H_c and W_h for Constant Oxygen

having a grain-size of 0.63 grains per square millimeter. With 0 per cent oxygen a single crystal sample containing 0.001 per cent carbon should have a maximum permeability of the order of 500,000 and a hysteresis loss of about 20 ergs.

For the sake of comparison with previous results[6] Fig. 13 has been prepared, showing the 1924 curve for hysteresis loss versus carbon content (broken curve), in which case oxygen was disregarded, and the present curves for 0 per cent and 0.02 per cent oxygen. The fact that the 1924 curve nearly coincides with the present curve for 0.02 per cent oxygen suggests that the 1924 samples with low carbon contained in the neighborhood of 0.02 per cent oxygen

[6]Loc. Cit.

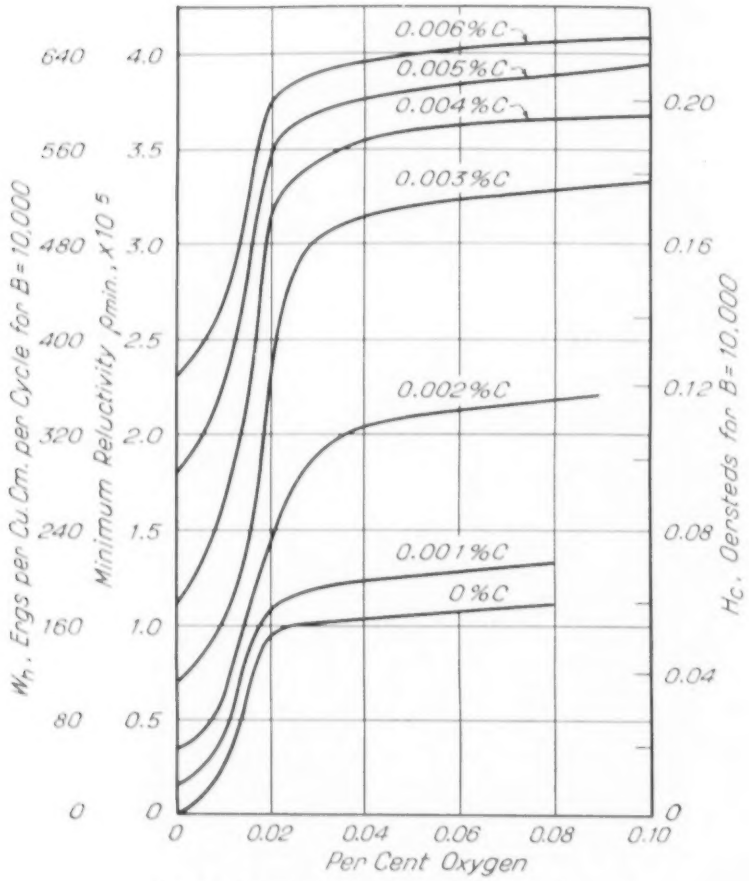

Fig. 8—Effect of O_2 on $\rho_{min.}$, Hc and Wh.

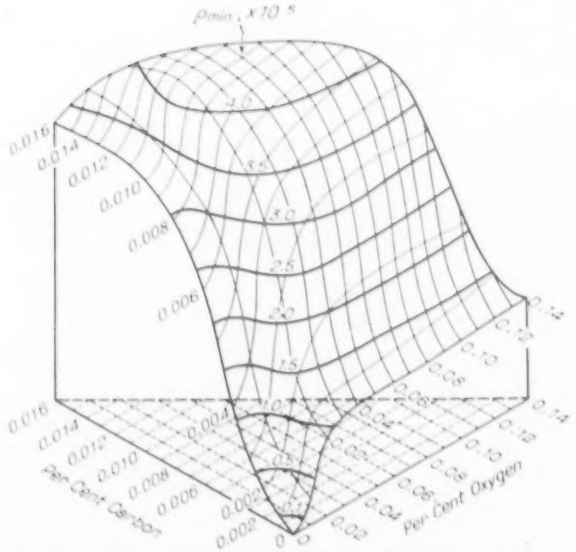

Fig. 9—Effect of Carbon and Oxygen on the Magnetic Properties of Iron. Numbers on Surface are Minimum Reluctivity $\times 10^5$ ($\rho_{min.}$). To Get Coercive Force (oersteds) for B = 10000 Multiply Numbers by 0.054. To get Hysteresis Loss (ergs/cc/cycle) for B = 10000 Multiply Numbers by 164.

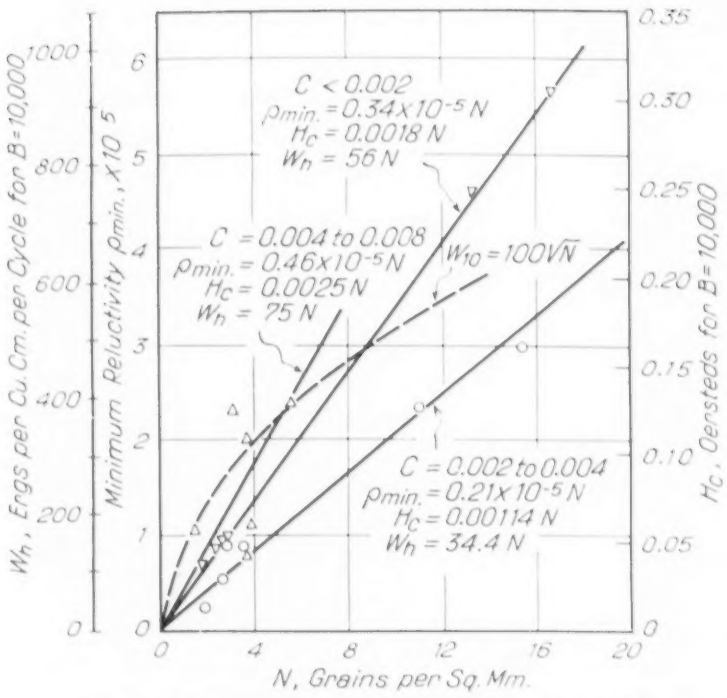
C < 0.002
ρmin. = 0.34x10⁻⁵ N
Hc = 0.0018 N
Wh = 56 N
C = 0.004 to 0.008
ρmin. = 0.46x10⁻⁵ N
Hc = 0.0025 N
Wh = 75 N
W10 = 100√N
C = 0.002 to 0.004
ρmin. = 0.21x10⁻⁵ N
Hc = 0.00114 N
Wh = 34.4 N
Wh, Ergs per Cu. Cm. per Cycle for B=10,000
Minimum Peluctivity ρmin., X10⁵
Hc, Oersteds for B=10,000
N, Grains per Sq. Mm.

Fig. 10—Net Effect of Grain Size on $\rho_{min.}$, H_c and W_h.

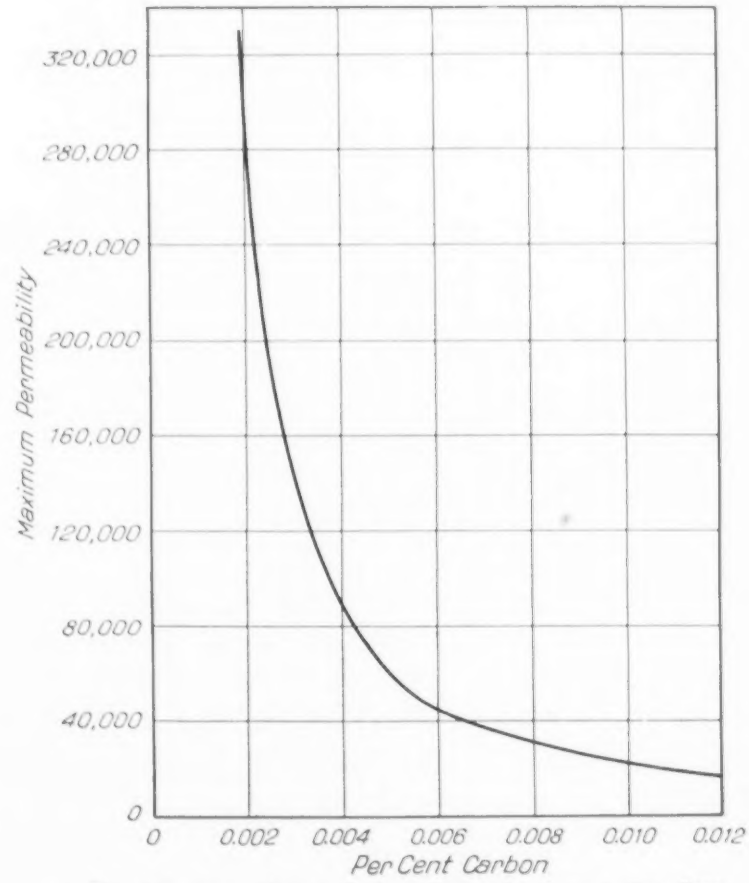

Fig. 11—$\mu_{max.}$ Versus Carbon Extrapolated to 0 Per Cent Oxygen and Single Crystal.

which is quite reasonable to assume. As the carbon content increases, the oxygen content becomes lower, and as the effect of oxygen also becomes less with increasing carbon content the curve for 0 per cent oxygen has been drawn in Fig. 13 so as to connect the end of the present 0 per cent oxygen curve with the 1924 curve at 0.08 per cent

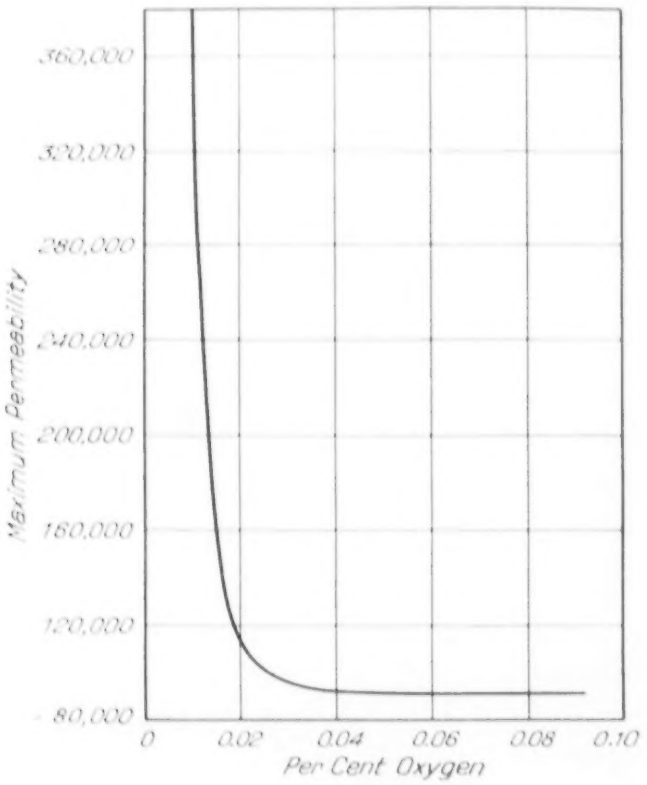

Fig. 12—$\mu_{max.}$ Versus Oxygen Extrapolated to 0 Per Cent Carbon and Single Crystal.

carbon. This then we regard as the net effect of carbon on the magnetic properties of iron. Coercive force, minimum reluctivity, and maximum permeability values for these higher values of carbon can be obtained by multiplying the hysteresis loss values by the proper factors, namely:

$$\rho_{min.} = 0.006 \times 10^{-6} W_h$$

$$\mu_{max.} = \frac{1}{\rho_{min.}} = \frac{164 \times 10^5}{W_h}$$

$$H_c = 0.00033\ W_h$$

From these various curves, (Figs. 9, 10 and 13) the magnetic proper-

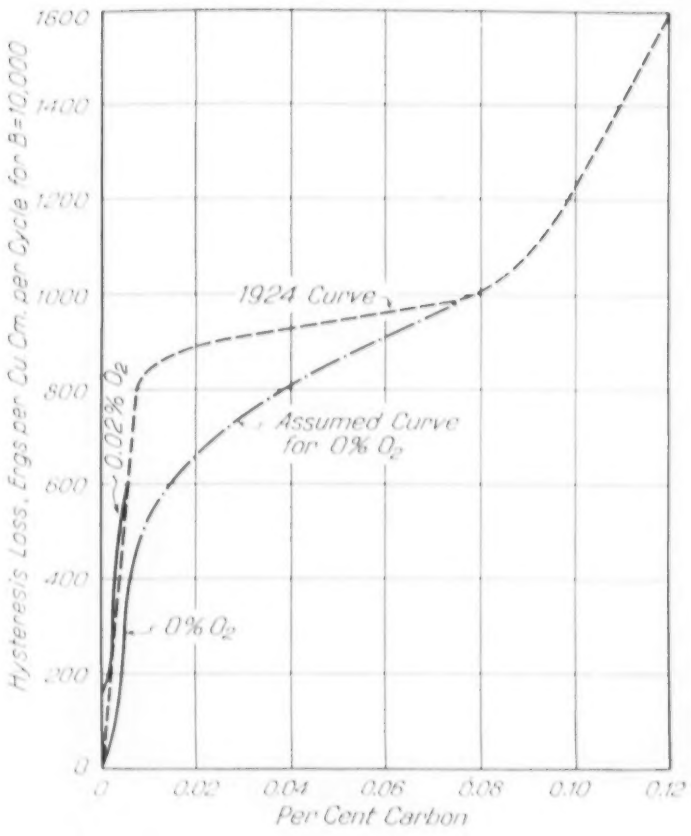

Fig. 13—Comparison with Previous Results.

ties of iron with carbon and oxygen each up to 0.12 per cent, and grain-size up to 20 grains per square millimeter can be determined with a fair degree of reliability. For small amounts of sulphur, manganese and phosphorus the previous results may still be used, namely:

$W_h = 18{,}000 \times$ sulphur (up to 0.06 per cent sulphur)
$W_h = 1{,}000 \times$ manganese (up to 0.6 per cent manganese)
$W_h = 13{,}000 \times$ phosphorus (up to 0.015 per cent phosphorus)

Conclusions

From Figs. 7, 8 and 9 it appears that the effect of oxygen on the magnetic properties of iron is very great up to about 0.02 per cent, from which point on the effect becomes inappreciable up to the limit of the present investigation, (about 0.15 per cent). The knee of the curves of Fig. 8 is indicative of a solubility limit at about 0.02 per cent oxygen. However, other evidence points to the solubility of

oxygen in iron at room temperature as being below 0.01 per cent[7] so that this point requires confirmation.

The effect of carbon is less than previously thought, as part of the effect attributed to carbon is now found to be due to oxygen. Even so, the effect of 0.001 per cent carbon is about ten times that of 0.001 per cent oxygen. The gradual change in curvature of the curves of Fig. 7 makes it more than ever doubtful whether there is a definite solubility limit at 0.006 per cent carbon as was concluded from the 1924 investigation. It is more probable that carbon—and perhaps also oxygen—is partly in solution and partly precipitated even at very low concentrations.

The effect of grain-size appears to be proportional to the grain number, N, and to vary with the amount of nonmetallic elements present. Although the shape of the curves differs from those of previous investigations the results show the same order of magnitude.

DISCUSSION

Written Discussion: By W. E. Ruder, Research Laboratory, General Electric Co., Schenectady, N. Y.

It is not surprising that data, other than that published by the authors in 1924, have not been forthcoming. The authors are to be congratulated on the patience and care involved in obtaining data on chemical analyses of impurities in this very low percentage range. The only question we can raise has to do with the validity of using an area for grain-size instead of a volume. Our reason for this is taken up more fully in a paper to be presented later in the week. In the same paper we call attention to the effect of orientation, a factor not heretofore considered, but one having an important effect on tests made on samples made up of grains of large size in relation to the total sample size.

O. W. Ellis[1]: I think it may be well to emphasize the tremendous amount of work that is represented by such an investigation, the results of which we have had presented to us by Dr. Yensen. I had the pleasure of seeing some of this work under way when I was with the Westinghouse Co. I think it can be said that this represents at least ten years of almost continuous labor.

There is one question I would like to ask Dr. Yensen in regard to this paper. Does he believe that he can determine, as a result of these experiments, the amounts of oxygen and of carbon that can enter into solid solution in iron, at room temperature?

N. A. Ziegler: The discussion of this interesting question, we were thinking of postponing until some future date, but I might say just a few words about it now. In parallel with the effect of oxygen on the low density perme-

[7] N. A. Ziegler, "Solubility of Oxygen in Solid Iron," Transactions, American Society for Steel Treating, Vol. 20, 1932, p. 73.

[1] Ontario Research Foundation, Toronto, Ontario.

[2] Metallurgist, Barber Colman Co., Rockford, Ill.

ability of iron we investigated the effect of oxygen on the high density permeability, and from the results (which are not included in this paper) I may say that oxygen in unalloyed iron has an effect on both.

According to our viewpoint, the low density magnetic properties—for example the low density permeability—are largely affected by impurities in solid solution or present as colloids whereas the high density permeability is affected by impurities present as precipitates. Since there is an effect of oxygen on both of these properties, I think we are justified in saying that in unalloyed iron oxygen is present in both forms. What the exact numerical values are of oxygen present as solid solution and as precipitated oxides, I cannot say, but there is a definite indication that it is present in both forms.

I do not want to go further into the discussion of this particular point because, as I said, it will be presented at a later date.

DR. T. D. YENSEN: I might add one remark to what Mr. Ziegler has said. If you look at the curve showing the relationship between oxygen and the magnetic properties, you will notice that the curve has a rather sharp knee between 0.01 per cent and 0.02 per cent oxygen, and it would not be unreasonable to assume that this knee represents the limit of solid solubility which then we may assume to be between 0.01 per cent and 0.02 per cent at room temperature. Some of you may have data to confirm or deny that statement. I would also like to refer you to the next to the last paragraph of the paper.

C. B. SADTLER[2]: I would like to ask Dr. Yensen if he has been able to determine any relationship between grain-size and minimum reluctivity when the grains are the result of fragmentation of the larger crystals, that is by cold working, when the sample that would be tested had been cold-worked and not heated above its recrystallization temperature.

DR. T. D. YENSEN: When we deal with magnetic properties, it is essential to anneal the sample thoroughly before testing. A test in the unannealed state is meaningless, except perhaps for high density properties. By annealing, of course, the sample recrystallizes and we get relatively large grains. The only answer I can give to this question is that what we have obtained is the relationship between grain-size in the final state and the magnetic properties, irrespective of how this grain structure was obtained.

DR. O. E. HARDER[3]: In comparing the 1924 curves with the present curve, the statement was made that oxygen apparently had less effect in those steels of higher carbon content. I wonder if there is a possibility that those steels of higher carbon content in 1924 were actually much lower in oxygen content, and if that is the factor involved in making those two curves more nearly coincide.

DR. T. D. YENSEN: I think Dr. Harder perhaps misunderstood what I meant. What I wanted to convey is that as the carbon content increases there probably is less and less oxygen in the sample. Consequently, for a carbon content of 0.1 per cent, we should expect the present results and the 1924 results to agree, much more nearly at least than they would for the lower carbon content in which we know that there is a certain amount of oxygen present.

As an apropos I would like to suggest that the term "steel" be confined to its proper sphere, that in dealing with very low carbon materials, very pure

[2]Assistant Director, Battelle Memorial Institute, Columbus, Ohio.
[3]METAL PROGRESS, Vol. 25, June 1934, p. 51.

materials, such as we do in magnetic materials, we would like to have the term "iron" used if possible. Would that be an unreasonable request?

G. C. McFARLAND: Where would you put the lower limit?

DR. T. D. YENSEN: Dr. Portevin* has suggested that we limit the term "iron" to the range in which the eutectoid transformation is lacking, and consequently in which there is no pearlite in the microstructure. In pure Fe-C alloys this will be the case if carbon is less than about 0.04 per cent. This is a high carbon content for magnetic materials of the type in which we are chiefly interested, and these should therefore all be referred to as irons. We must go below 0.01 per cent in order to get good magnetic properties.

Steel, Dr. Portevin suggests, should be confined to the range from 0.04 per cent to 1.7 per cent, and cast iron to alloys with more than this amount of carbon.

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